METALLURGIA

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(INCORPORATING THE METALLURGICAL ENGINEER)

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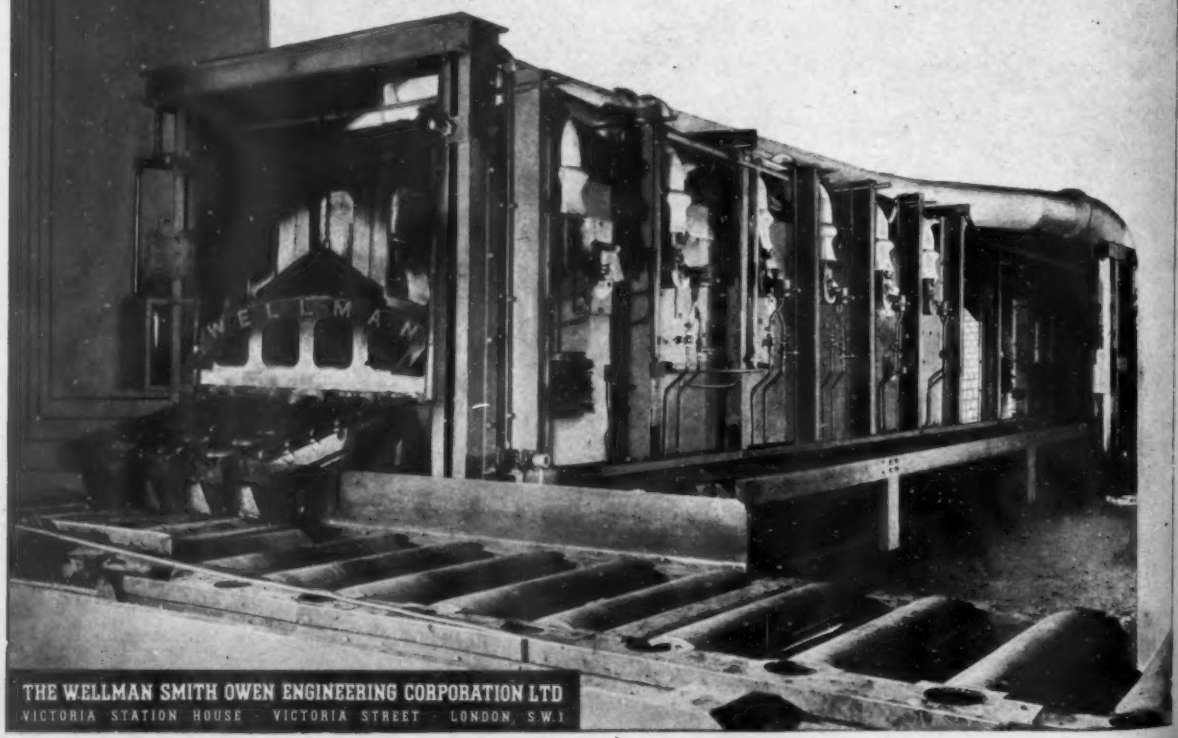

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METALLURGIA

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World Mineral Supply Outlook

AT world power conferences, geological congresses, and the like, where men of science and their more or less scientific technical brethren foregather, impressive and rather gloomy prognostications of early exhaustion of vital mineral wealth are uttered especially of the wealth associated with power supply—coal and petroleum; although it is a curious fact that, so far as the geologists are concerned, with each succeeding conference the estimated reserves, say, of coal, are greatly augmented. Thus, after ten or twenty years of rapidly-growing consumption, the remaining reserves like the widow's oil, have increased, or not diminished.

But it is only natural that during a period of prolonged war, when consumption of essential minerals leaps up on a prodigious scale, and the spectre of scarcity appears in many supply fields, there should be some apprehension as to the future; sharpened by real or apparent anxiety about future generations which also characterises a war period. We are told that minerals, unlike vegetable material, do not grow: once consumed, they are gone beyond recall or replacement. Some indeed "grow" in a geological sense, but it is a slow business involving millions of years; and for the practical purposes of that second of cosmic time representing, say, 4,000 years, this form of replacement or increase can be ignored.

We learn that the vast deposits of rich iron ore in the Lake Superior region are being rapidly depleted. Mr. Shinwell has just informed us that some of Britain's best coal reserves are nearing exhaustion; and as for petroleum, that should have been all used up long ago, but with almost uncanny persistency it bubbles up, so to speak, in various parts of the earth with all the joy and energy of eternal youth. Petroleum, knowing little and caring less about the much-vaunted atomic energy, intends to drive the engines and oil the wheels of industry and commerce for many years to come; even if Senator H. Cabot Lodge, of Massachusetts, in his recent melancholy list of early demises among U.S. minerals, gives American oil only another fifteen years of life. It appears, too, that iron, steel, copper, nickel, chromium, tin, zinc, lead, will soon be no more, and must early submit to the indignity of a suitable epitaph—on a plastic tombstone.

The Senator certainly seems in a hurry to get rid of much of America's mineral wealth, and bring in the ungilded age of plastics and nuclear power. For example, he limits the life of bauxite supplies (for aluminium) to three years; of tungsten, mercury, lead and zinc to five years; of copper to 10 years; petroleum 15 years; and iron ore 20 years. There seems to be a fatalistic flavour about the five-year term and its multiples, or at least, the unexpected precision of arithmetical progression. Mr. Shinwell says we are very short of skilled mining engineers. The prospects in that profession would, however, seem to be a little grim; especially as the view has been expressed that the above forecasts might, with a few exceptions, apply to the whole earth.

But there are extenuating circumstances and gleams of hope, and some time must elapse before Campbell's "Last Man" makes his impassioned harangue to the dying sun. If one considers only the relatively rich deposits and the present or recent abnormal rate of consumption, the outlook does not appear bright; but if the potential supplies and alternatives, together with scientific and technical progress, are brought into proper perspective, apparent present scarcity may eventually turn into abundance. Consider only the results of some recent prospecting and geological survey: there are now very large proved reserves of copper in Rhodesia and Chile; manganese in Russia and India; iron ore in Brazil and Sweden; and petroleum in Venezuela, Iran, Iraq and Arabia. There are, of course, vast areas in China, Siberia, and other parts of the world, hardly yet explored, still less thoroughly prospected. A conservative estimate of petroleum resources, for example, can now regard these as sufficient for another hundred years, even though this may mean drilling to greater depths than the present limit of about three miles. Coal is mined to a greater depth. In fact, a recent survey has suggested the possibility of regarding the earth's crust available for mineral supplies as ten miles thick.

Before considering what this means in terms of potential supplies of various minerals, it may be worth while to look at some alternative sources of power, to ease the demand for coal and oil. There is, in the first place, little need to emphasise the supreme economic importance of water power, from rivers and ocean tides: white coal, as it has been well named. It is perhaps stretching a point to say that this can grow or increase like materials of vegetable origin; yet it is certainly related more or less directly to vegetative growth. Water power depends on volume of water flowing in the rivers; that depends on rainfall; and that in turn depends to a considerable extent at least on forests and afforestation in the river basin. Power from the tides, as a practicable proposition, is probably a little more remote, belonging to the more distant category of sun-power and atomic power, although with 25 to 30 ft. tides, such as are common around Britain, it is surprising that this form of power has not been applied.

It may well be argued that some of these latter categories are no more remote or impracticable than delving ten miles into the earth's crust. Why not, indeed, go a little deeper and utilise the heat of the supposed molten interior of the globe for power and other purposes? But this is venturing too far into the

realms of Jules Verne; though, who is there so rash as to assign a limit to the possibilities of science and technology, including the nightmares of H. G. Wells?

The earth's crust, including the hydrosphere (ocean), and the lithosphere (land), contains vast quantities of the four minerals: coal, iron, aluminium and magnesium, also of course silicon. Based on a U.S. geological survey, the percentage composition of the crust is: silicon 27·72, aluminium 8·13, iron 5·01, calcium 3·63, sodium 2·85, potassium 2·6, magnesium 2·09, titanium 0·63, manganese 0·1, barium 0·05, chromium 0·037, copper 0·01, tungsten 0·005, zinc 0·004, lead 0·002, cobalt, boron and beryllium 0·001, molybdenum 0·0001, mercury 0·00001, silver 0·000001, arsenic, tin, and gold 0·0000001. This last percentage, or one ten-millionth part of the earth's crust, represents 24,500,000,000 tons of gold, or arsenic, or tin; and from this the astronomical figures for the other and more plentiful constituents may be calculated. These estimates look impressive, but the financier, if not also the mining engineer, will not be impressed, and will merely enquire: How much of it is payable?

Into the intricate questions of mineralogical economics it is not possible to enter here. It may well be that, long before mineral supplies are approaching exhaustion, financiers will have been abolished, together with their awkward scepticism.

But on the technical side there is much ground for hope. The latest estimates for coal, of a practical and conservative kind, ensure a supply for at least 4,000 years. It's more economical use, in which there is room for vast improvement—e.g., in hydrogenation and production of liquid fuel, will make a tremendous difference in the outlook. In the economical use of low-grade iron and other ores remarkable progress has been made in recent years, and the end in that direction is by no means in sight. The production of many new alloys of the lighter metals, aluminium and magnesium, has wrought a revolution in our own day in many ways, including economy of iron and steel, of copper and other metals.

It is fairly obvious that complete exhaustion of supply of any one mineral will not be sudden, but very slow and gradual, thus giving mankind plenty of time and opportunity to make the necessary adjustments, to find substitutes, and generally to adapt himself to new materials and methods. He has had to do this throughout the whole of his existence on this planet, and is not likely to lose the art. And there are always plastics.

The Royal Aircraft Establishment

BEFORE presenting prizes to the students of the Royal Aircraft Establishment Technical College, at Farnborough, recently, Mr. Arthur Woodburn, Parliamentary Secretary of the Ministry of Aircraft Production, emphasised the need for scientific and technical training. In asserting that British science saved the world from Hitlerism, he stated that the liberty of the world in 1940 depended on the speed and manœuvrability of British fighting aircraft and their aeronautical superiority in the Battle of Britain, and on the development of radar which provided our defenders with an all-seeing scientific eye. Until it is possible, continued Mr. Woodburn, to rely on the security of International Law, it will, unfortunately, be necessary to provide for the eventuality of war. The lessons of the past indicate that nations which lagged only slightly behind in the development of weapons, tactics, strategy and scientific thought found themselves outfought, outmanœuvred, and defeated.

Germany had done considerable scientific work in preparation for the war, but had failed to give her theories practical application. The Royal Aircraft Establishment was recognised and appreciated by aircraft manufacturers as the laboratory for the solution of fundamental aeronautical and engineering problems, and it was the close connection between the scientists and manufacturers which had secured the maximum efficiency with the minimum of waste. Of the students at the College, Mr. Woodburn said they were entering a new world of discovery and achievement, and the College was giving an example in the combination of general education with high technical qualifications. Students are encouraged to realise the implications of their work and its effect on the community, and education for citizenship is a vital part of the curriculum.

This country was badly let down at the beginning of the war by the scarcity of directive ability in all grades of industry. What we require to-day is quality which provides the basis on which Britain will be able to give service to the world. It would be discouraging to think that our training and education were only serviceable in war, but nearly every great stride in the technical achievements of warfare has brought immense benefits in civilian life. With the advent of atomic energy the lesson should be plain to us all—that a nation which remains self-centred and complacent, and which places its reliance upon a proud and obsolete insularity, is heading for extinction.

In conclusion, Mr. Woodburn referred to several examples of recent contributions which the Royal Aircraft Establishment has made to aeronautics, and said that, by the work of aeronautical science, the world was now entering upon an air age, and, while we have achieved what, to our forefathers, would have been regarded as miracles, really we were only on the threshold of discoveries difficult to appreciate.

The British Diamond Die Federation

A NEW association, known as the British Diamond Die Federation has been formed. It is composed of suppliers of raw stones (rough diamonds), manufacturers of diamond drawing dies and users of diamond dies. Its formation will enable the various members of this industry to continue the co-operation which was commenced in 1940 under the ægis of the Diamond Die Control. During the war the efforts of all concerned were concentrated on building up a new industry in this country capable of supplying drawing dies to replace those previously imported from abroad. These efforts met with conspicuous success.

It will be the duty of the new Federation to ensure that this industry, which was introduced into this country in times of war, is continued with equal success in peacetime. The industry is capable at present of meeting all the home demand for dies. It will be the Federation's job to see whether it can be expanded so that it can also contribute to the national export trade and whether it can continue with equal success at home in competition with foreign suppliers. The new Federation will carry on much of this work, which was initiated by the Diamond Die Control, and its formation received every encouragement from the Control.

Bursting Defects in Cast Aluminium Cylinder Heads

By S. A. J. Sage

(*Late Metallurgist, Armstrong Siddeley Motors, Ltd.*)

Some problems were encountered in the production of cylinder units for an air-cooled cylinder engine. The unit was composed, mainly, of a cast aluminium alloy head, screwed and shrunk on to a forged and machined steel barrel and one of the problems concerned small bursting defects on the surface of the cast heads which were discovered on removal from the solution temperature salt bath. The investigation, described in this article, was carried out to determine the cause and to effect a solution.

PRIOR to, and during the heavy bombing period of 1941, it was the national policy to disperse high priority work from the concentrated, specialised areas to less industrialised parts of the country. Many firms which had previously built complete units, now, through expansion and dispersion of sub-assemblies found themselves as mother factories with many small sub-contracting or shadow additions. At first, these small additions, often with only a very minute nucleus of skilled craftsmen, were somewhat helpless and dependent on the mother factory. But, as time went on, these small factories became very highly specialised in the production of the particular work they were carrying out. They became more and more reliant on their own efforts and less burdensome on the parent organisation. This does not mean to say that the greater experience of the parent factory was ignored, but rather it was incorporated and modified to suit the local conditions and new experiences gained by the operatives concentrating on the sub-assemblies.

One of these factories was given the task of producing cylinder units for an air-cooled radial engine. The unit was composed basically of a cast aluminium alloy head, screwed and shrunk on to a forged and machined steel barrel. The present investigation is concerned with one of the problems which occurred, for a period, on the cylinder head. This problem concerned small bursting defects on the surfaces of heads which were discovered following solution treatment of the castings.

At the time when the trouble arose the heads were being cast by three suppliers, each of which, whilst casting under fairly closely controlled conditions, had their own peculiar methods. At that time it was hardly compatible to hold raw material suppliers to definite methods of casting production. However, of these suppliers, two did their own heat-treatment, but the other, through lack of furnace equipment, sent the fettled castings to the cylinder dispersal factory for heat-treatment. It was this situation which made the tracing of the bursting defect more difficult. The heat-

Fig. 1.—Several bursts occurring on the rocker-box and flange of a cylinder head after solution treatment in a salt bath.

Fig. 2.—Bursts in the rocker-box and top fins of another head illustrating the "crater-like" appearance of the defect.

Fig. 3.—More bursts in the top fins of another head.

treated castings of the other two suppliers were received in good condition.

The analysis of the cylinder head material was :—

Copper	1·3 to 2·5%
Nickel	0·5 „ 1·7%
Iron	0·8 „ 1·3%
Magnesium	0·7 „ 1·7%
Silicon	maximum 2·0%
Titanium	maximum 0·25%
Aluminium	remainder.

The heat-treatment given to the heads consisted of solution treating by heating to 520°–530° C. for 3 hours followed by air cooling in an oven fitted with internal fans. This gave an even, reproducible rate of cooling. Ageing was then carried out by heating to 190°–200° C. for 5 hours in a gas-fired air circulating furnace followed by air cooling.

The bursting defect was first found when one of the batches of heads was removed from the solution temperature salt bath. Two heads out of the batch of 24 had small craters in the rocker boxes and fins as shown in Fig. 1. These heads were cooled in the cooling chamber and when cold thoroughly examined. It was found that the general appearance of the casting was not good but was inclined to be "scabby." The large crater in the side of the rocker-box was situated on the parting line of the mould of the casting. Due to the undercut nature of the head the mould is made in two halves with the rocker-boxes vertical. When the two halves are put together they are sometimes out of register, so the resultant casting shows the parting line of the mould. This particular line, when fettled, causes no loss in the efficient working of the head but it can always be seen if the head is examined carefully.

Around the edges of the crater, from its appearance it looked like the remains of a small volcano, a white powder appeared. The amount of this increased as the head was exposed to the air.

As many thousand heads had been treated in the existing salt baths prior to the finding of these two heads, production continued whilst the heads were submitted for examination. However, the next batches of heads were free from the bursting defect.

Micro-examination of the heads which had burst showed that the bursting was connected with large inclusions embedded in the metal. The metal adjacent to the bursts showed signs of slight overheating, but areas from projecting portions of the heads, which should be the first to show signs of overheating if the salt bath had been at too high a temperature, showed no such signs. The inclusions were metallic and, from the appearance of the surrounding structure, appeared to be able to dissolve aluminium but not the $CuAl_2$ hardeners which were present in the structure. This can be seen in Fig. 4, the matrix of the inclusion being the dark half tone and the hardeners the light areas. The powder which formed at the edge of the bursts was white and brittle but on prolonged exposure deliquescence occurred. This reaction seemed to be continuous because the liquor reacted with surrounding aluminium to form more white powder.

During the succeeding weeks the number of heads bursting increased. The bursts were found in almost every position on the head although at first they seemed to be connected with those parts of the casting which would be to the top of the mould during the casting operation. Over a period, 86 heads had burst in the salt bath out of a total of 1,608 treated. Two salt baths had previously been used but correlation between burst heads and specific salt bath was bad.

Heat-treatment Procedure

At this stage the complete heat-treatment procedure was overhauled to try to isolate the heat-treatment from the casting operation.

The solution treatment was carried out in salt baths approximately 4 ft. × 3 ft. × 3 ft. high. The tanks were of mild steel sunk partly into the ground with 6 "pot" type gas burners playing on the bottom. They were fitted with a heavy cover through which two thermocouples in sheaths were placed. The salt was of the 90% sodium nitrate—10% potassium nitrite type. The heads were placed rocker box up in a two-stage welded mild steel cradle which held 6 heads in the bottom stage and 6 in the top stage. Two of these cradles filled a tank. The cradles were handled by a chain-driven set of pulley blocks running on an overhead runway which led from the salt bath tanks to the cooling oven. The procedure was to thoroughly clean and dry the heads, load them and their respective test pieces into the cradles and suspend them over the salt bath for preheating. The salt bath temperature was adjusted to 460° C. The cradles were then slowly lowered into the bath, the lid closed, the thermocouples re-inserted and the temperature raised to 520° C. The heads were then soaked for 3 hours after reaching the solution temperature. The heating control was by hand but the limits 520°–530° C. were not exceeded. The couples were replaced periodically with calibrated couples checked against a master couple and potentiometer. It was therefore known that the couples would show the temperature of the bath within 1° C. at 520° C. The recording instruments were checked against a standard potentiometer and couple. The only variable temperature gradient could have been from any hot spot occurring in the bath. The bath was then explored at many positions to find out whether there was any variation in temperature between the salt temperature and that being recorded by the two instruments already installed. No hot spots and very little difference in temperature was found in either of the two baths between a standard couple and potentiometer and the two recording instruments. The error was of the order of 2° C. which was found near the surface of the salt in the centre position of one of the baths. This was only found when the bath was heated to temperature without a load of heads. No difference existed when the bath was loaded.

To make doubly sure that the two original baths were not at fault a third was made up, using new salt, checked thermocouples and recorders. Batches of heads were again found to be defective after solution treating.

After making certain that the equipment was satisfactory the procedure was next examined. The workmen carrying out the heat-treatment had already satisfactorily processed many thousands of heads, and after checking their methods it was ascertained that the technical instructions of the parent company were being rigidly followed.

Mechanical Tests

Mechanical tests taken on many hundreds of heads, both defective and good, showed no differentiation either in hardness or maximum stress—both being extremely uniform, even from cast to cast.

As the defective bursts occurred in the solution treatment, the remainder of the heat-treatment schedule and

Fig. 4.—X 100. Shows how the aluminium matrix has been "attacked" by inclusions leaving the hardeners in a relatively unchanged condition.

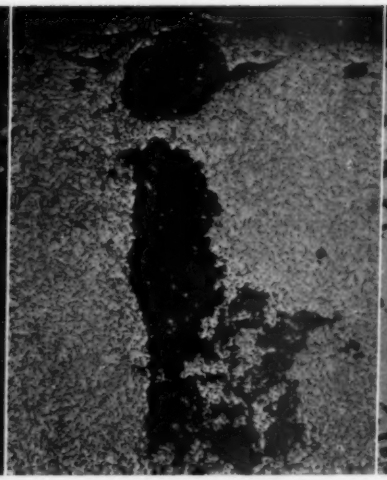

Fig. 5.—X 20. Shows an inclusion which has burst the surface of the casting. This was observed on the surface as only a small "pimple."

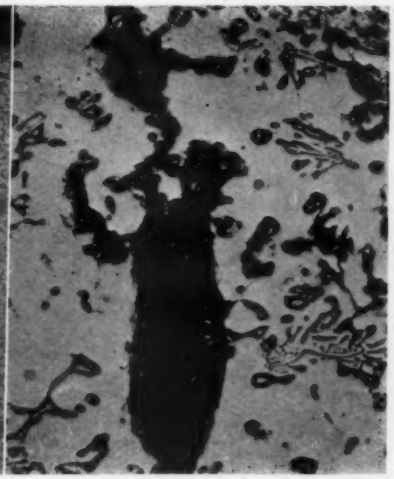

Fig. 6.—X 100. Shows a large inclusion surrounded by streamers of flux.

equipment was not examined so thoroughly, but even these showed no deviations from the standards previously set.

Statistical Study of Defective Heads

The defective heads were next studied statistically to try to find out whether they were all coming from one melting furnace, one moulder or out of one heat-treating bath. Figures taken over a period showed that out of 3,400 heads heat-treated, 155 had burst, this necessitating the quarantining of 712 heads. The 712 was composed of 77 casts or melts, 42 of which contained 10 or more heads per melt. The majority of these melts had only one or two burst heads per melt, but on nine occasions five or more bursts were found in one melt. However, of these nine occasions eight were from batches containing more than 10 heads per melt and the other occasion was from a smaller batch. Again, there was very little, if any, connection between the cast number and the burst heads; i.e., there was no method of forecasting that if one batch showed a poor result the batch numbers prior to or following that number also would be poor. Thus there seemed to be no correlation between the furnace and the defects; i.e., the defects were found from all furnaces.

The code numbers which correspond to individual moulders were also checked, but, apart from the fact that two moulders out of seven had more rejects than the others, all moulders were susceptible to giving heads which would eventually turn out to be faulty.

The melts and batches of heads were next studied in relation to the heat-treatment baths. It was soon shown that whether a batch was confined to one bath or divided among the three baths the defects were entirely connected with individual heads and not with melts or batches.

Examination of Head Sections

Whilst these statistics were being drawn up further heads were sectioned and examined. Sections were invariably taken from defective areas, but in some instances further sections were taken from portions of the head away from the defects; or again from a head which had been heat-treated at the same time as one showing defects, but which had not itself shown any defect.

These sections again confirmed the original finding—that the bursts were connected with inclusions—but they also showed that the inclusions were almost entirely due to flux and dross entrapped during solidification of the aluminium alloy in the mould.

Section (a) taken from the flash line of the exhaust spring box, but not cutting through blisters or bursts, showed two small isolated inclusions. Due to expansion—but not to bursting—one of these just below the surface had produced a small bead on the latter. The grain size was very fine and showed no variation between the top and the bottom of the section. The microstructure was normal for that section and the entrapped dross had had no effect other than that of bursting the surface. Until this defect was sectioned it had not been in contact with air but on exposure it started to react and form the white powder mentioned previously.

Section (b) cut through two defects in the inlet spring box. The defects appeared to be small on the surface but the sections showed them to be $\frac{1}{4}$ in. to $\frac{5}{8}$ in. deep respectively. They were filled with inclusion matter which was extremely hard and unaffected by either etching acids or 2% NaOH. The inclusions in these cases were surrounded by streamers of dross which had reacted with the adjacent aluminium matrix but not with the hardeners.

Section (c) cutting through one of the large bursts seen in Fig. 1 showed no signs of inclusions but in view of the violent reaction which has taken place this was not unexpected. Streamers of dross extending well into the section were however, still to be seen.

Section (d) was taken from a head which had been heat-treated but showed no signs of bursting. The section was cut from the spring box along the flash line.

No signs of any large inclusions were found but a few small inclusions were noticed. In each case the inclusions were surrounded by streamers but in no

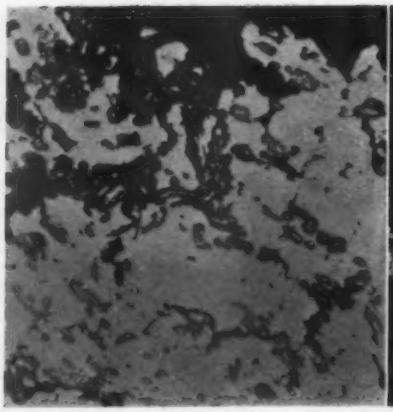

Fig. 7.—X 100. Shows the dross streamers at the bottom of a burst.

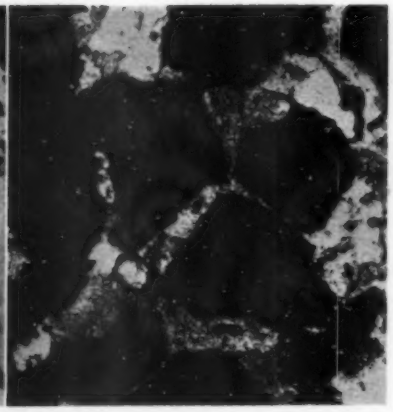

Fig. 9.—X 100. Shows inclusions, mainly of Al-Ni-Fe, which have not burst. It shows their hard and brittle nature.

instance were either the inclusions or the streamers in contact with the surface of the casting. The microstructure at the bottom of the spring box showed a preponderance of Mg_2Si hardeners but no further inclusions or bursts were found at this portion of the head.

Section (e) was cut from the top fin which adjoins the two rocker boxes. As it showed many bursts it was examined to see whether it would give any evidence of inclusions trying to "float" as they should do when this portion of the head is being cast. The microstructure was similar to that of a previous section in that segregation of Mg_2Si and NiAl was present, the Mg_2Si being more preponderant. Many inclusions and streamers were found, the inclusions being intact but the streamers showed signs of having reacted and in some cases were still reacting whilst on the microscope stage. This took the form of small bubbles coming from an interaction between the streamers, which were deliquescent, and the surrounding aluminium. This seemed to be a continuous reaction, white powder being formed which grew more extensive as time went on. No variation in grain size was found.

Section (f) was cut from the thick base of the casting in a front position and included two bursts. The latter showed the white powder as soon as the section was prepared, this continuing to develop as the defect was exposed to the air. This section of the casting would be a high spot in the mould and it is thought that this was the case of the segregation of NiAl, the dross, and the inclusions which in turn have caused the bursts. The top edge of the section was very porous and contained many inclusions and reacted streamers. The reaction seemed to progress along the NiAl hardeners but these were left unreacted. The hardeners of this micro section were generally widely spaced but tended to segregate approximately half way down the section.

Section (g) was cut from a rocker box intersecting three bursts and several "pimples." One burst remained inactive after sectioning but the other continued to react. All the reaction, however, was in the streamers which surrounded the inclusion, the latter generally remaining inactive. Again the microstructure showed segregation of NiAl and Mg_2Si at the base of the rocker box. As this section was still reacting it was decided in order to prove a theory, to reheat-treat the section. This was done and the result showed that the streamers, which were deliquescent, had reacted violently in the salt bath and burst the section. The large inclusions which were surrounded by the streamers reacted with the salt and increased in size. This was also confirmed with many other minute inclusions; they all increased in size.

Section (h) was cut from a fin showing streamers and inclusions. Reheat-treating this section again showed that the streamers reacted violently and that the inclusions gained in size. The microstructure away from the reacted area was quite normal and the grain size was small. The structure adjacent to the streamers showed signs of overheating.

Section (i) was taken from the crown of the cylinder head at approximately the centre position. The burst in this section showed definite signs of inclusions. The microstructure showed that the majority of the hardeners were of the "chunky" type—probably due to the special cooling of this area of the casting. The surface of the crown, although rough machined prior to heat-treatment, showed fairly extensive porosity to a depth of 0·050 in. A few large inclusions were found at the bottom of isolated holes.

From these sections the following conclusions were reached :—

1. That heads showing bursts are found after heat-treatment.
2. That different heads from the same batch or melt show variations when heated in the same heat-treatment bath.
3. That different batches in the same bath give variations after heat-treatment.
4. That bursts occur at any position on the head but usually in an area that was at the top of the mould during casting.
5. In every burst examined the residue consists of a large inclusion of hard unetchable Ni-Al-Fe compound surrounded by deliquescent streamers of flux from the casting stage. The streamers react when exposed to the air and if reheat-treated in this condition form new bursts.
6. The whole inclusion and the streamer do eventually react by forming a deliquescent liquid which in turn reacts with the surrounding aluminium.
7. That inclusions of Ni-Fe-Al burst the surface, without reacting, merely by expansion. If the inclusion is far enough from the surface of the casting the expansion seems to be taken up by the surrounding metal thereby causing areas of weakness.
8. That bursting only occurs when streamers of flux are present.

Suggested Theory of Cause of Defect

The theory of the mechanism of bursting and its cause is thought to be :—

When a cylinder head is cast, inclusions of flux and dross occur. If this flux and dross is completely

surrounded by good material, with no connection with the surface of the casting, they only expand during the normal heating in the salt bath. However, if the inclusions are near the surface of the casting the latter is pushed outwards to accommodate the expansion of the inclusion. If the flux and dross reach the surface of the casting, either by their original position in the casting or by the upheaval of the surface due to the expansion of the inclusions, another reaction, much more violent, occurs. The nitrates and nitrites of the salt bath react with the fluorides of the flux with the result that the whole area of flux and inclusion is blown outwards from the surface of the casting.

Casting Operations Studied

In view of the above findings which absolved the heat-treatment from blame, attention was then paid to the casting operation at the raw material suppliers. The following facts were ascertained :—

Melting was being carried out in two types of furnace. One, the larger, was a coke-fired bale-out crucible type furnace capable of melting up to 500 lb. of aluminium alloy. The other was of the oil-fired tilting type melting up to 250 lb. The large furnace produced enough metal for a batch of ten heads and over whilst the smaller ones took care of the smaller batches of heads.

The procedure was to heat the aluminium alloy rapidly to 700° C. and cover the metal with flux to the extent of 4 ozs. per 100 lb. of alloy. The metal was then degassed and after a specified time skimmed and poured or baled into two crucibles. These were then skimmed again and the metal poured into the moulds.

To cover all angles of the casting technique, samples of floor sand were examined and compared with that of another supplier whose heads were showing no defects. Visual examination, sieve analysis and clay content were studied ; also compressive strength and permeability.

The results obtained were :—

Visual examination : Both samples appeared normal when compared with that of the " satisfactory " supplier.

Sieve analysis gave :—

Sample (a) from " satisfactory " supplier.
Sample (b) and (c) from " defective " supplier.

	B.S.S. Sieve No. and % retained.									
	16	30	44	72	100	120	150	200	240	240
Sample (a)	0·4	0·4	3·0	12·0	16·4	9·0	12·0	6·4	13·0	14·0
Sample (b)	0·8	0·8	1·8	9·2	16·0	8·0	12·4	7·0	14·0	18·0
Sample (c)	Tr.	Tr.	0·5	8·0	17·0	9·0	11·0	8·0	18·0	17·0

Clay content (by washing).

Sample (a)	14·0%
Sample (b)	12·5%
Sample (c)	11·0%

Compression and Permeability :—

	Green.			Dry.	
	Moisture %	Comp. lbs. sq. in.	Perm.	Comp. lbs. sq. in.	Perm.
Sample (a)	7·8	4·45	34	56·0	38
Sample (b)	7·25	4·45	17	43	34
Sample (c)	7·75	4·3	16·3	46	28

Except for the lower permeability of the sand from samples (b) and (c) as compared with those of sample (a) there is very little difference between the sands used. The results are normal for this type of unmilled sand.

Attention was then turned to the fluxing of the molten alloy. It was found that with the flux then in use, the scum and oxides formed were of a fluid type and whilst every effort was made to effectively skim and prevent this from entering the mould with some of the molten metal, some obviously escaped. It was thought that this was the main cause of the bursting trouble. Another supplier, who was not having the bursting trouble, stated that they were using a different type of flux at the rate of 8 ozs. per 100 lb. This flux formed a dry powdery dross on the top of the melt which was easily separated from the molten metal. Although the first flux was being satisfactorily used on other types of castings in somewhat similar alloys the second flux was tried using the same melting technique but with 8 ozs. of flux per 100 lb. of melt.

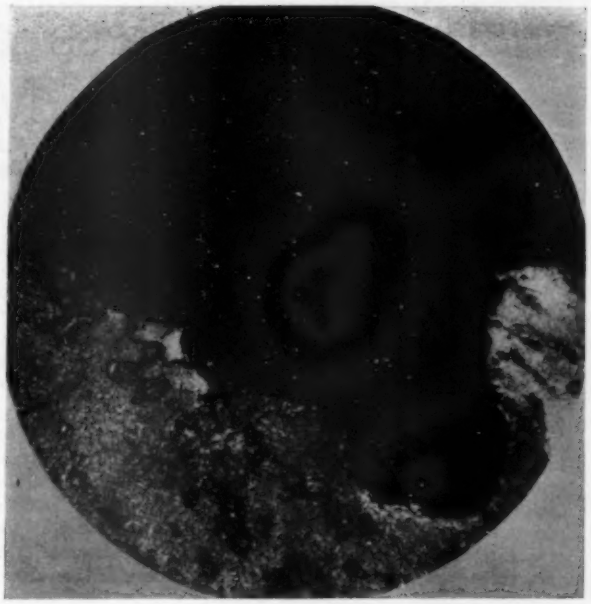

Fig. 8.—X 100. Shows an inclusion at the bottom of a crater. This had not re-acted but it may have pushed the surface out.

Samples were cast, fettled and heat-treated in the same salt baths as those in which all heads had previously been treated. The results were satisfactory, no bursting defects being found. Mechanical test figures, hardness and machineability were in no way affected by the change in flux. The foundry stated that the change over to the new type of flux entailed no difficulties and since then over 500 heads have been cast and heat-treated without any re-occurrence of the bursting defect.

Conclusions

That the bursts found on the surface of an aluminium alloy were due to entrapped flux which reacted with the hot salts of the heat-treatment bath. In view of the fact that no defects were found on the surface of the casting prior to heat-treatment it is considered that the reaction had to be started by the disturbance of the surface of the casting due to a relatively high expansion of large Ni-Fe-Al inclusions near the surface which allowed the salt to come into intimate contact with the fluoride flux. Where these inclusions were deeply set in the metal no contact was made between the salt and the streamers of flux and the surrounding aluminium seemed to be able to accommodate the expansion of the inclusions. By changing from a flux which forms a fluid dross to one which forms a powdery dross the difficulty was overcome. The powdery flux is easier to separate and inclusions of reactive nature are eliminated.

Acknowledgment

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Internal Stress and the Fracture of Metals

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Methods of testing such as are called for in purchasing specifications may be excellent in so far as they act as a check on the quality of the material, but, when the material is shaped or built up to form a component, many other factors influence the capacity of the material to resist fracture. One common cause of fracture is due to unsatisfactory conditions of internal stress in a component and in this paper, recently read before the Ebbw Vale Metallurgical Society, the author discusses some aspects of the relation between internal stress and the fracture of metals, and also suggests how the risk of failure may be diminished or avoided.

CONTINUITY of production in industry depends on numerous factors, and one of these is unfailing operation of plant and machinery. A common cause of breakdown is the fracture of a machine part, and such fractures are produced by unsatisfactory conditions of internal stress in the parts. The serious import of the relation between internal stress and breakdown is vividly indicated in the disastrous collapse of the first Quebec Bridge in 1907, when 80 workmen lost their lives. This catastrophe was brought about by the development of internal stresses which could not be safely carried by the metal forming some of the bottom chords of the bridge. The purpose of this paper is to discuss some aspects of the relation between internal stress and the fracture of metals, and also to suggest how the risk of failure may be diminished or eliminated.

It is unnecessary to dwell on the importance of good quality material and a high degree of homogeneity of the mass of metal comprising a machine part. Defective material requires no discussion to prove its unsuitability for vitally important mechanical operations in industry. It is assumed, therefore, that the metal is in a good physical condition, and free from any defects to which fracture could be attributed.

The Effect of Internal Stress

Metals are crystalline solids possessing an atomic configuration which may be elastically distorted and partially or completely disrupted by internal stresses. The limit of disruption is fracture of the metal, and there is, therefore, between the limit of elastic distortion and fracture, a dangerous transition stage which may be encroached upon when the internal stresses exceed certain critical values. These critical values are, broadly speaking, the elastic limit and the endurance limit of the metal, and each is related to a dangerous transition stage which depends on how the internal stresses are produced. In the case of the elastic limit, the transition stage is represented by the near approach to, and the occurrence of, plastic yielding, and it is associated with stresses produced by dead loads and occasional non-cyclic loading. In the case of the endurance limit, the transition stage is represented by the initially minute but progressive disruption of the original atomic configuration at isolated points, and it is associated with repeated stresses produced by cyclic loading, or by some other form of loading combined with motion of the machine part—usually rotary motion.

Those engaged in industry have ample evidence around them in the form of fractured parts to show that the transition stage associated with repeated or cyclic stressing is the more important. This is so, for the further reason that the progressive disruption, in this transition stage, towards fracture, is not accompanied by any easily observed external warning. On the other hand, failure under high steady stress, for a wide range of materials, is preceded by a considerable amount of plastic deformation which cannot easily escape notice. The important requirement, therefore, for avoiding fracture in machine parts is to ensure, by careful design and controlled operation, that the recurring internal stresses at points of high stress concentration are never allowed rise to values above the endurance limit of the material. The most irritating feature of so many fatigue fractures is, that while almost the whole of the metal comprising a machine part—perhaps 99·99%—is initially, never stressed above the endurance limit, an isolated high recurring stress, operating on a minute part of the metal, is sufficient to initiate progressive cracking and fracture of the machine part in service. Other important aspects of this matter will be discussed at a later stage.

The Causes of Internal Stress

Internal stress is mainly brought about by external influences, and these may be classified as follows :—

(i) The application or extraction of heat, in association with resisted expansion or contraction, respectively.

(ii) The application of forces which may produce, separately or in combination, direct loading, bending moments or twisting moments.

(iii) The motion of mass at high speed, and in particular, the vibration of mass at high frequency.

These three main causes of internal stress may operate separately or in various combinations, and very serious internal stresses may arise from the resultant effect of stresses of the same kind, produced at a point by the separate causes acting together. Each of the causes is considered in more detail below.

(i) Internal Stresses due to Thermal Changes.

In the case of internal stresses produced by thermal changes, it is important to observe that the constraint on thermal expansion or contraction may be provided within the mass of the machine part itself. In hard and non-ductile materials, a thermal condition may produce internal stresses which would fracture the metal with explosive violence. A good illustration of this is provided by the effects of applying heat at too fast a rate to chilled cast iron rolls. In order to consider these effects, let us assume that one of these rolls is initially cold and at a uniform tem-

perature throughout. When hot steel is being rolled continuously, there is a corresponding input of heat into the roll until a fair degree of thermal equilibrium is established, and this initial heating is relied on to bring the rolling face into correct shape. In order to accelerate this process, flame-heating is sometimes resorted to for a short time before hot steel is passed between the rolls. It is generally recognised that this initial heating must be carried out with care, and the reasons for this are explained below.

During the period of initial heating, an examination of the temperature variation along a radius of the roll at successive points of time would show results approximately like that shown in Fig. 1. The numbers on the curves represent successive points of time. The slight cooling effects when the roll

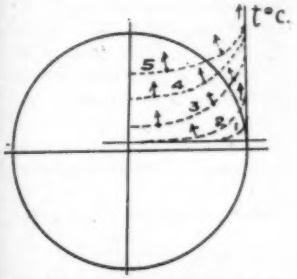
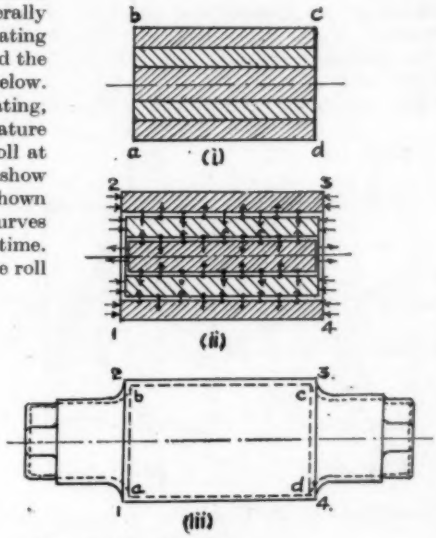


Fig. 1.—General form of the radial temperature variation during the heating up of a roll.

surface is not in contact with the hot steel are omitted for the sake of simplicity.

The effects of the differences of temperature can be understood more clearly if we imagine that the roll, in its cold, uniform temperature condition, is composed of a large number of tubes of diminishing size surrounding a central core. It is assumed that each tube and the core is free to expand in accordance with the thermal change within it. Fig. 2, (i).

The outer tube, having the highest temperature, would have the greatest expansion in both an axial and a radial direction, and there would be lesser expansions of a similar kind in the other tubes and the core. An exaggerated representation of the changes produced is shown in Fig. 2, (ii). In the solid roll, free expansion of the different zones cannot occur, and therefore, there must be a considerable amount of varying strain with corresponding tensile and compressive stresses. The distribution and the nature of these stresses may be gauged from the imaginary external forces which would have to be applied to the different zones to strain them sufficiently to restore the mass of metal to the form taken by the solid roll in accordance with its thermal condition. Fig. 2 (iii). It will be observed that these restoring forces show :—(1) That in the outermost zone, compressive



(i) Zoned roll—cold.
(ii) Zoned roll—heated.
(iii) Initial and final form of the solid roll.

Fig. 2.—Effects of differences of temperature on a roll.

stresses are set up in the axial and the circumferential directions, and tensile stress in a radial direction ; (2) that in the inner core, tensile stresses are set up in all three of the principal directions. The intensity of these stresses will depend on the temperature gradient, and this gradient depends on the rate of heat input. If the heating of the roll is too rapid, then the roll may be fractured in the manner shown in Fig. 3†.

The Fig. shows a chilled cast-iron tinplate roll, 28 in. diameter × 36 in. long (body) × 22 in. diameter, necks. It was a new roll, and it had been put into the housings in readiness for commencing work on a Monday morning. A little time before the beginning of the shift, the roll was being preheated by an oil-flame. Just within one hour after preheating began, the roll suddenly fractured transversely and longitudinally, as shown in the photograph. The longitudinal crack was visible on each side of the roll, but it did not pass completely through the roll neck. The original photograph was marked in ink to show the line of the crack. On the basis of the earlier discussion, there is reason to believe that heating had been too rapid, and in view of the much greater strength of cast iron in compression than in tension, the thermal conditions set up would subject increasingly large internal transverse and longitudinal areas to tensile stress and produce internal fracture and almost instantaneous extension of the fracture to the outer zones of the roll.

Fig. 3.—A chilled cast-iron roll broken by too rapid heating with an oil-flame.

In this matter of the stresses produced by thermal changes, it is impor-

† From the author's paper, "Failure of Chromium-Molybdenum Rolls." *Proc. Inst. of Mech. Engrs.*, Vol. 128, 1934.

tant to note, in passing, the serious internal stresses which may be produced during the casting and the heat-treatment of machine components. With good understanding and control of thermal changes in metals, the residual stresses may be largely eliminated or diminished to small values without serious effect on the service life of the machine part. But, control is difficult or inadequate sometimes, and the possibility of incipient cracking by the action of high internal stress must be carefully considered. For important parts, it is highly desirable that modern crack detection methods should be employed to prevent those parts which are cracked, passing into service. These methods also make it possible to exercise watchful control of cracked parts which may be put into service.

It is necessary to lay equal emphasis on the need for careful control of thermal changes throughout the whole life of a machine part. The insidious effects of thermal changes are easily overlooked because these changes cannot be seen, except indirectly with the aid of suitable instruments, and when we have knowledge of the changes, it is not easy to estimate their full effect in producing internal stress. It is useful to remember that thermal stress effects are greatest and most harmful when the rates of heating or cooling are highest. Also, wherever possible, freedom should be provided for expansion and contraction, and where complete freedom cannot be provided, it is sometimes possible to so design the part that the stresses due to temperature changes are kept to low values. Special care is required when there is a possibility of a severe state of thermal strain coinciding with the application of heavy loading. The internal stresses produced by mechanical loading are considered in the following section.

(ii) Internal Stresses Produced by Applied Loads.

When a load P, Fig. 4a, is steadily applied to a bar of metal of uniform cross-section, there is reason to believe that if the stress values were measured on exceedingly small areas, they would be found to vary in an irregular manner across a section of the bar according to the equilibrium of the force system on the crystal structure within each minute area considered. When the stress is referred to the customary larger unit area—viz., 1 sq. in. or 1 sq. cm., the stress has an almost uniform value across the section, as shown in Fig. 4b. But, this condition may be expected only when the resultant line of action of the load coincides with the axis of the bar. When the line of action is eccentric, as shown in Fig. 4c, the internal stress varies from a maximum on one side of the bar to a minimum on the other side. When the line of action of P reverts to the axis of the bar, but with the uniformity of section interrupted by the presence of a hole, the type of stress variation is that shown in Fig. 4d. If the load P and the area of the cross-section at XX in each case is the same and equal to A, the stress per unit area—i.e., P/A is only true for the bar shown in Fig. 4b. It is the average stress in the other cases, and while designers make use of average stress values, it is the maximum stress which they always seek to evaluate. The foregoing stress variation has been discussed only in connection with steady loading, and it is important that the effect of other modes of applying the load P should be considered.

Dynamic Loading.—The manner in which a load is applied to a machine part has a corresponding effect on the magnitude and variation of the internal stress. We are concerned here with the non-cyclic type of loading, which includes a wide variety of occasional impact and shock loading. With this type of loading, the elastic distortion of mill plant parts is much greater, and because the internal stress is proportional to this distortion, high values of these stresses may be produced momentarily, becoming for moderate shock, two to four times greater than the stress produced by the same load when it is gradually applied. Under such conditions, permanent deformation or fracture of parts may occur unless adequate provision has been made against the effects of these high stresses.

The estimation of steady loads in industrial service is not difficult, and design for satisfactory service can be carried out with a high degree of certainty. The estimation of dynamic loading which characterises so much of industrial plant operation, is more difficult, and it is often desirable to verify calculated load conditions by making direct measurement on plant parts. Uncertainty here, can itself become the initial cause of fracture and consequent production delay. There is also special need for care when plant operators may unwittingly overload or abuse machinery.

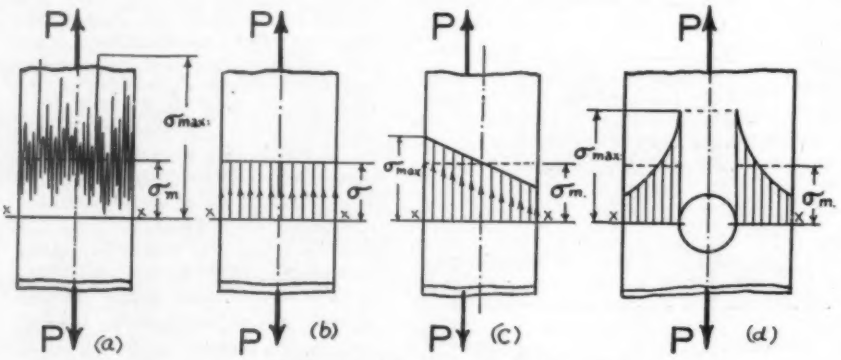


Fig. 4.—Apparent and real stress.

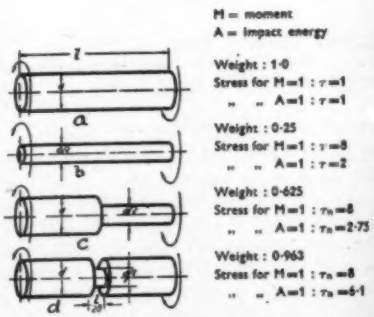


Fig. 5.—Torsional impact stressing in relation to design.

In addition to accurate evaluation of dynamic loading, it is necessary to consider the strain energy capacity of a machine part. An unfavourable disposition of the whole mass of metal comprising a part may so lower the strain energy capacity that serious deformation or fracture may occur. A simple and interesting comparison illustrating the general effect of dy-

namic loading is shown in Fig. 5, taken from a paper by Thum[1]. Several bars of steel of the same length, but different in form are subjected to the same amount of torsional impact energy A, and this is momentarily taken up as strain energy in each bar. If the momentary maximum shear stress T which is produced in the bar *a* is 1 (unity), then the corresponding maximum shear stresses in the bars *b*, *c* and *d* would be represented by the figures 2, 2·75 and 6·1, respectively. The high value in the case of the bar *d* is brought about by the lower total resilience of the small diameter part, when compared with the bars *b* and *c*. The maximum shear stresses for a static twisting moment M are also shown.

the mode of elastic flexure, estimation of the internal stress variation is difficult. It is necessary in such cases to carry out careful experimental measurements of stress on machine components under operating conditions. This form of stress investigation has developed extensively in recent years, and this has been made easier by the concurrent progress in electronics. The resistance of a metal to failure under cyclic stressing is largely independent of the manner in which the stresses are produced, and it is for this reason that the experimental measurement of stress is so important. When this can be supplemented by mathematical analysis, progress towards the elimination of fractures can be more rapid and certain.

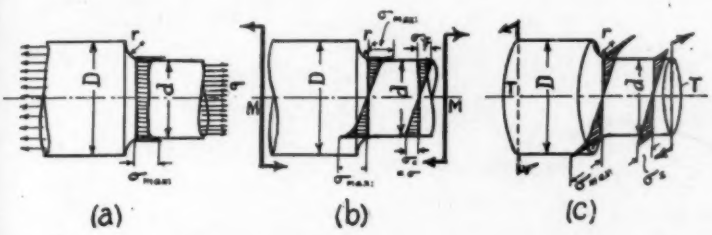


Fig. 6.—Showing the effect of fillet radius on the stress concentration.

N.B.—In all cases, the value of k depends on the values $\frac{d}{D}$ and $\frac{r}{d}$

Approx. ranges for the values of k :—
(*a*) 1.3—3.0 (*b*) 1.1—2.6 (*c*) 1.1—3.2

(*iii*) *Internal Stresses due to the Motion of Mass at High Speed.*

Increase in the speed of motion of machine and engine parts has been an outstanding development in engineering progress during the last few decades. It has given rise to many new problems associated with internal stresses in metals. The stresses are produced by inertia forces which may be static in their effect or cyclical in character, as in the case of vibration. The estimation of these forces is bound up with the analysis discussed under dynamic loading in the last section. The same inertia forces may be produced by large masses moving at low speeds as with small masses moving at high speeds. These inertia forces may not be associated with vibration, and their effects in producing internal stress can be calculated in many simple cases with a great degree of accuracy. But, when vibration of high frequency occurs, additional stresses, varying in a cyclical manner, will arise, and since these depend on

Other methods of measuring internal stresses have been developed, and chief among these are: two and three dimensional photo elasticity; the application of brittle lacquers to the surfaces of machine parts; the strain-rosette and X-ray diffraction.

The Effect of External Form on Internal Stress

One of the most serious causes of high internal stress is a sudden change of the surface profile at some part of a machine member which is already subjected to a high average stress. A high percentage of the service fractures that occur are located at places where there are sudden changes of section—e.g., at the shoulder fillets of shafts, at the root of screw threads, under the heads of bolts and rivets, etc. The production of high local stresses under such conditions has been comprehensively examined by some of the experimental methods referred to[2], and the general character of the internal stress variations which occur is shown in Fig. 6.

In Fig. 6a σ is the uniform direct stress over the small diameter cross-section due to the external load. The maximum stress σ at the fillet is $k\sigma$, where k varies from 1·3–3, as the radius of the fillet becomes smaller.

In Fig. 6b the shouldered shaft is subjected to a bending moment M which produces a maximum tensile stress σ_t and a maximum compressive stress σ_c in the smaller section. At the fillet, the stresses reach maximum values $= k\ \sigma_t$ where k varies from 1·1–2·6, as the fillet radius becomes smaller.

Likewise, in Fig. 6c, when the shaft is under a torque T, the maximum shear stresses occur at the fillet, and have the value $= k\ \sigma_s$, where σ_s is the maximum shear stress in the smaller section and k varies in value from 1·1–3·2, as the fillet radius becomes smaller.

Cyclic Stressing

The foregoing discussion has been mainly concerned with the factors which produce and alter the values of the internal stresses. Earlier in the paper reference was made to the serious effects of cyclic stressing. Internal stresses, which may be safely carried when they remain steady or are only occasionally applied, become dangerous when they are applied in a cyclic manner. Under this mode of stressing, when the internal stresses are at just above the safe limit, there is, according to Gough and Wood[3],

Fig. 7.—A fatigue fracture at the shoulder fillet of a hollow shaft.

a progressive disintegration of the crystal structure in the vicinity of the high stress, and this process eventually reaches the stage at which an exceedingly fine crack is formed. The boundary of the crack gradually

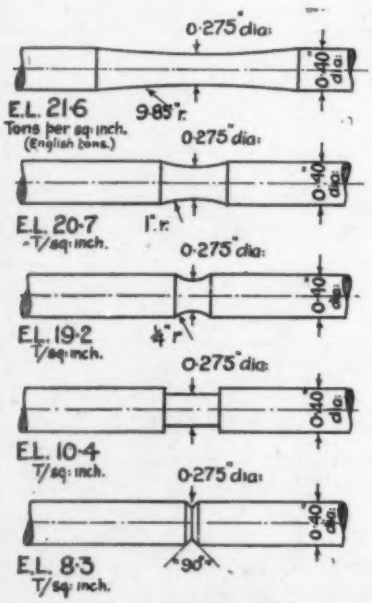
[1] *Proc. International Engineering Congress, Glasgow*, 1938.

[2] "Design Data," *Amer. Soc. Mech. Eng.*, March, 1939.

[3] "Deformation and Fracture of Mild Steel under Cyclic Stress in relation to Crystalline Structure", *Proc. Inst. of Mech. Eng.*, Vol. 141, 1939.

spreads into the metal until the remaining section of the part is unable to resist complete tearing, giving the typical form of complete fracture shown in Fig. 7.

The maximum value of the repeated stress in tension and compression when the mean stress is zero is called the endurance limit, and the determination of this limit for different metals is one of the important objectives in fatigue testing. The procedure, in this form of testing, is to subject a number of specimens of the same material to different known values of stress, either by direct loading, as in the Haigh machine, or by



E.L. = Endurance limit.

Fig. 8.—The effect of change of shape on the endurance limit. (Moore and Kommers).

the application of bending moments, as in the Wöhler machine or the two-point loading rotary bending machine. In the case of the latter machines, the specimens are rotated, to produce repetition of the applied stress. Specimens which are unable to resist the effects of the repeated stresses fracture in due course, and the number of cycles of repeated stress to fracture corresponds to the number of revolutions made by the specimen before fracture. By plotting the number of cycles to fracture against the stress which produced fracture, and also those particular cases in which fracture does not occur even after a very large number of cycles, an endurance curve is obtained. This curve would show that with decreasing values of the applied stress, the number of cycles before fracture occurs increases, and finally a stress value—viz., the endurance limit, is obtained which just fails to produce failure of the material by fatigue. For ductile materials, the endurance limit is about 0·4–0·5 of the ultimate tensile strength of the material.

Factors which affect Fatigue Resistance

Unfortunately resistance to fatigue, found in the manner described, has not the uniqueness that belongs to the static tensile strength and the elastic limit. There is a vast amount of experimental data which shows how resistance to fatigue failure is affected by numerous factors which have little or no effect on the ultimate tensile strength and the elastic limit of the material. Some of these factors are :—the quality of the surface finish ; the shape of the machine part ; corrosion ; understressing and overstressing—i.e., controlled application of certain assigned values of cyclic stresses before a part is put into service. On the matter of surface finish, fatigue resistance may be increased by shot peening the surface or by surface rolling under high pressure. The harmful effects of corrosion may be eliminated by applying non-corrodible

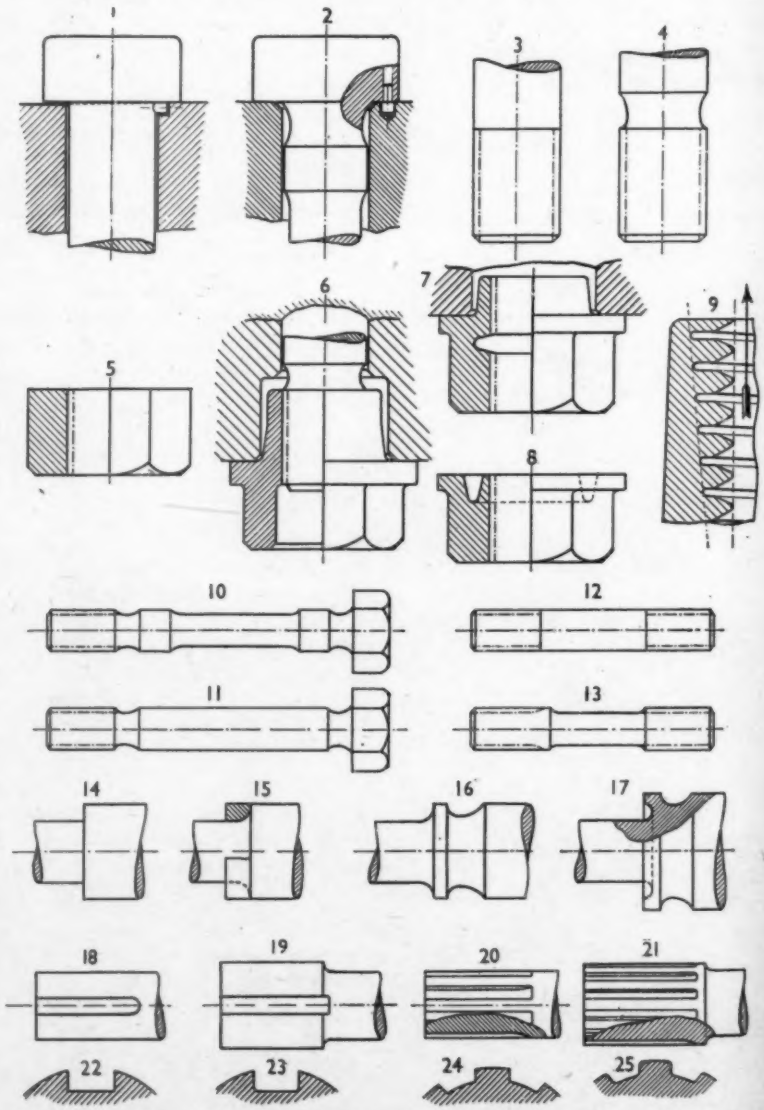


Fig. 9.—Defective and improved design of parts subjected to variable stress.

metal coatings. A striking indication of the effect of surface profile is shown in Fig. 8 which summarises work carried out by Moore and Kommers.[4] Although the smallest cross-section is the same in each case, the altered forms of the surface profile produced the different values of the endurance limits.

[4] "Fatigue of Metals," by Moore and Kommers. Published by McGraw Hill.

In view of the considerable effect of different factors on resistance to fatigue failure, it is not easy in design to ensure that internal stresses will never reach dangerous values. But, progress in the removal of doubt and chance by careful experimental and analytical processes has been very remarkable during the last twenty years. There are innumerable examples of machines and engines that call forth much admiration for the work of the engineer and the metallurgist. In addition to successful adaptation of machines and engines to the operations of industry and commerce, the machine designer has succeeded in a large measure in bringing about complete continuity of operation. But continuity of operation also depends on the general attitude of mind of industrial operatives. They cannot be expected to visualise the effects of varying applied load on the internal stresses, and it is a good and fairly common practice to provide limit stops and limit marks to guide their manual operations. In the loading of machine parts, the variation of the loads is always accompanied by corresponding average variations of internal stress. Local stress concentrations do not always vary in the same manner as the loads, and it is for this reason that special care is needed in design. Machinery is designed for certain assigned conditions, and freedom from breakdown may often be achieved when every care is taken by management and operatives to comply with those conditions.

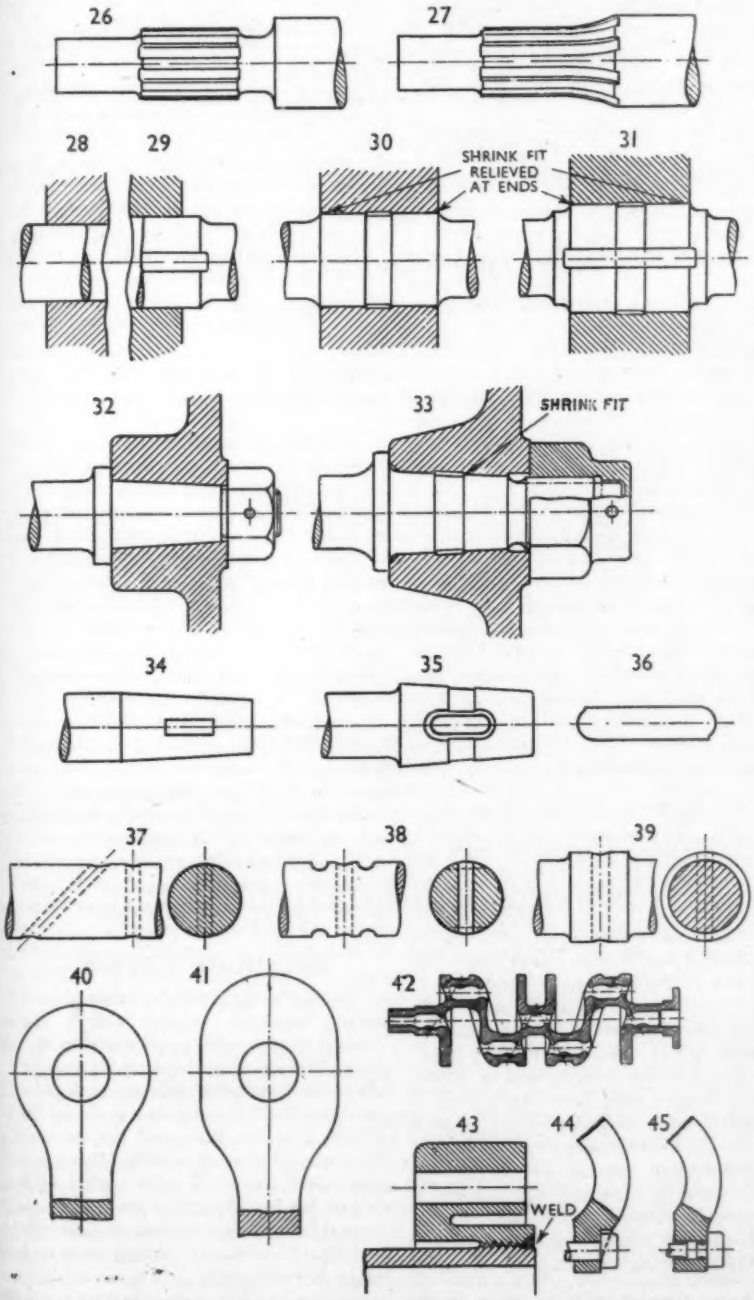


Fig. 10.—Defective and improved design of parts subjected to variable stress.

Good and Bad Design

In order to crystallise some of the points discussed in the paper a number of examples of good and bad design are shown in Figures. 9 and 10 and the following explanatory notes are directly associated with them.*

Bolt Heads.

†(1) Defective form of bolt head and incorrect location of the snug or dowel. Also defective seating of the head.

(2) Improved form of bolt head and better location of the snug, as suggested by Schuster (1933).[5]

Bolt Ends.

(3) Defective form of bolt end. Change of section from thread to shank is too sharp.

(4) Improved form of bolt end. The change from threads to the shank is less abrupt, and stress concentration is much reduced.

Nuts.

(5) Defective form of nut.

(6) Improved form of nut giving tensile and compressive stresses within the body of the nut. Described by Volk (1938).[6]

(7) and (8) Improved forms of nut for relieving stress in threaded part of bolts.

* Taken from the author's paper, "Research on Materials and Engineering Design." Published by *Inst. of Mech. Eng.* (South Wales Branch), 1939.

† The figures refer to the numbered sketches in Figs. 9 and 10.

(9) Method of stress relieving in large nuts. Described by Volk (1938).[6]

Shock and Creep.

(10) Good form of bolt for shock resistance. Higher capacity for storing strain energy than a parallel shank bolt.

(11) Good form of bolt for resistance to creep at elevated temperatures. Smooth change of section, and the large shank diameter gives lower stresses and thus lower creep rate than a bolt with a reduced shank diameter.

Studs.

(12) Defective form of stud.

(13) Improved form of stud.

Shoulders.

(14) Defective form of shoulder.

(15) Form providing for a square shoulder without abrupt change of section.

(16) and (17) Stress-relieving grooves for shoulders. Described by Burn (MacGregor, Burn and Bacon 1935).[8]

Keyways and Splines.

(18) and (19) Defective and improved arrangement respectively for a keyway.

(20) and (21) Defective and improved arrangement respectively for splines.

(22) and (23) Defective and improved form of keyway section. Provision of small radii in the improved form.

(24) and (25) Defective and improved spline sections.

(26) and (27) Defective and improved form, respectively, of a spline shaft. Described by Lehr (1937).[7]

Shrink and Press Fits.

(28) and (29) Defective forms of shrink and press fits. The seat should not project beyond the hub or nave.

(30) Improved form when no key used.

(31) Improved form when a key is used. The hub or the seat should be machined to the very slightest taper at the ends so that the pressure here is less than that over the remainder of the hub and seat. Note that shoulders are kept flush with the faces of the hub.

Tapered Ends and Cotter Holes.

(32) Defective fitting of tapered end of a rod. Changes of section are too sharp.

(33) Improved form of fitting. Boss tapered. Change of section less abrupt. Thread shielded from corroding influences.

(34) Defective form of cotter hole. The square corners are sources of weakness.

(35) Improved form of cotter hole. Mortising carried out by milling and not by drilling and filing.

(36) Sharp corners produced when cotter hole is drilled and finished by chipping and filing.

Holes in Shafts.

(37) Shaft weakened by drilled holes.

(38) and (39) Form of stress relieving described by Burn (MacGregor, Burn and Bacon, 1935).[8]

Miscellaneous.

(40) and (41) Fig. 41 shows an improved form of Fig. 40. Described by Coker (1931).[9]

(42) Cast crankshaft of suitable design. Described by Thum (1938).[10]

(43) Improved patented form of pipe flange for high pressure and temperature. Designed to relieve stress on the pipe walls. Referred to by Bailey (1934).[11]

(44) and (45) Fig. 44 shows defective form produced by counterboring. In the improved form under-cutting is avoided.

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New Zinc-Conductor Alloy

By W. Deisinger and R. Reinbach

BESIDES copper and aluminium aldrey (0·4–0·5 Mg, 0·4–0·6 Si, 0·2–0·4% Fe, remainder Al), zinc, zinc alloy Zn-All (made of pure Zn 99·99% with 0·7–0·9% Al and 0·35–0·5% Cu) have been used as conductors for power transmission.

In the last few years, a Zn-Fe alloy with 0·13% Fe has gained prominence in Germany. It has a tensile strength 16·5–17·5 kg./sq. mm. (23,400–24,900 p.s.i.), an elongation at fracture of 25–40%, and a bending number of 25–35. The electric conductivity is 15·5–15·8 mho/m. sq. mm. The recrystallisation, which is high in Zn-All, is much lower in the new Zn-Fe alloy.

A comparison shows :—

Condition	Zn-Fe	Zn-All
Annealed at 100° C. . .	0·1	0·5
Cold-deformed	0·15	0·4
Annealed at 150° C. . .	0·3	0·8
Annealed at 200° C. . .	0·5	0·9
Annealed at 250° C. . .	0·7	0·98

Complete recrystalli ation would be indicated by 1 ; it shows that the drawn Zn-Fe wire is just beginning to recrystallise, while Zn-All is already half recrystallised. The commercial material has a zinc content of 0·10–0·20%. It behaves in casting, therefore, almost like pure zinc—i.e., it solidifies with coarse dendritic crystals if no means are taken to prevent this.

Extrusion casting proved to be successful in obtaining finer and better directed crystals in chill mould castings, which gave a material without tendency to rolling cracks. As this alloy has a lower tensile strength and higher elongation than Zn-All, this quality has to be taken into account in drawing.

The first two or three passes should be made with reductions of 10–15%, the later passes with about 20–25%. To draw a wire from 8–1·8 mm. diam. requires, therefore, about 15 passes.

The tensile strength and elongation as function of reduction is, contrary to Zn-All, practically constant—namely, 18 kg./sq. mm. (25,600 p.s.i.) and 40–43% respectively between 10 and 80% reduction. The tensile strength corresponds about to that of the semi-hard aluminium, while elongation at fracture lies considerably higher. The slightly lower conductivity is of no practical importance.

As the recrystallisation temperature lies very high, no intermediary annealing need be introduced between operations since the creep strength is not affected by the cold-deformation.

Metallwirtschaft, 1944, **12** pp., 56–59, from *Materials and Methods*, 1945, **22**, 1478.

Beryllium Copper

Although not used in large quantities, beryllium copper alloys are finding considerable application in the manufacture of springs and in sheet form for instrument diaphragms. In the commercial compositions there is 1·9–2·25% bery lium, with the balance copper, except where small additions of not more than 0·5% of nickel and cobalt are introduced to prevent grain growth. Beryllium copper springs are extremely resistant to shock, wear and fatigue.

From *Con. Met. Metallurgical Ind.*, Oct., 1945.

Indium Plating Aids Drawing of Aluminium Alloys

By E. E. Halls

Investigations have shown that an electroplated coating of indium has lubricative properties and helps considerably in the drawing of aluminium alloys. Although indium is one of the expensive metals it is logical that this metal should be used when the benefits derived merit its use. These notes refer more particularly to the electro-deposition of indium as practised commercially in America.

ONE of the advantages of using indium as a plating for dies is its value as a lubricant. The use of indium for this purpose is based on the investigation of the lubricative properties of their metal films by F. B. Bowden and D. Tabor. Tests were made by users on various dies indium plated for trial purposes. Reports showed that for drawing aluminium, the electro-plated indium coating aided considerably. The work was carried out on a double crank, open back, mechanical draw press, and the aluminium alloy employed was Alclad 24 SO. A summary of the conclusions reached are as under :—

(*a*) With indium plating, a job which would normally require two or three draws, could be completed in one draw.

(*b*) With indium plating, quicker overall operation enabled two or three components to be produced in the time normally required to lubricate the tool and produce one part on the same machine.

(*c*) The general increase in the life of dies plated by the indium process proved to be of the order of 50%, and the following specific instances are cited :—

(i) In one case, the life of striking dies was increased from seven to eight minutes, to three hours.

(ii) In a second case, the life of the tools was increased from that for 10,000 operations to that for 18,000 operations.

Indium is, of course, one of the expensive metals, and a special technique for its electro-deposition has to be acquired, although once this is mastered and process conditions are controlled, the plating procedure is simple and no difficulties need arise. Two main processes have been commercialised in America, and have been established for some time, having given invaluable service in war applications. It is logical that specialised applications where the benefits derived are realistic enough to merit its employment, should now be sought. As with chromium plating, it is no doubt a question of the right type of plating properly applied to suit specific needs, rather than a general and indiscriminate introduction of the method.

In consequence of the foregoing, a few notes on the electro-deposition of indium as industrially practised in America, are of interest. Two processes have been patented, one using a cyanide electrolyte by the Indium Corporation of America, the other, employing a sulphate solution, by the American Smelting and Refining Corporation. They are operated to process instructions on the basis of the purchase of the materials or proprietary compositions.

Indium itself is a silvery white metal, malleable and ductile, rather softer than lead, with a low melting point of 155° C. It is stable in dry air, but corrodes rapidly in moist air. At first sight, this makes it success for wear and corrosion resistant coatings rather surprising. Actually, its plating involves a duplex coating and entails a heat-treatment operation, and this probably explains the apparent discrepancy in full. Appreciable flexibility exists in the choice of the undercoat, but the general indication is that the lower the melting point of the metal used for the primary layer, the better are the results. Heat-treated indium coatings on lead or tin are claimed to be superior to those on silver or copper, and cadmium is the basis extensively employed. For specific purposes, lead, tin, copper, silver and gold have all been used in addition to cadmium.

The heat-treatment properly executed causes thorough diffusion of the indium into the primary coating. It is commenced below the melting point of 155° C. to promote the diffusion of the indium into the base metal and to avoid its melting and globulising on the surface. It is completed a little above the melting point, 175°–180° C. being the usual maximum, the time period of heat-treatment being determined by the thickness of the indium coating. An oil bath, hot-air oven or controlled atmosphere furnace can be used, but full temperature control should be provided.

The preliminary procedures for electro-plating follow along orthodox lines—i.e., solvent degreasing for the major oil or grease, electrolytic alkaline cleaning for final traces of contamination, acid pickling for rust, if necessary, on ferrous metals, and bright acid dipping for tarnish on non-ferrous materials.

The " finish " of the final indium coating is determined partly by grinding, lapping and polishing the base metal, and partly by a light buffing of the soft plating after heat-treatment.

The primary plate is applied in the usual way—viz., cadmium from a cyanide electrolyte, lead from a fluosilicate solution, and copper, gold and silver from cyanide electrolytes. The particular undercoat applied must be very thoroughly washed in cold and hot waters before passing to the indium solution, this being essential to avoid contamination of the latter. Further, the work must not be allowed to dry off before entering the indium electrolyte.

The Indium Corporation's electrolyte is made to the following nominal composition :—

Indium (as indium chloride)	4 oz. troy.
Sodium Cyanide (free)	12 „ „
Dextrine	2 „ „
Water	1 gal. (American).

The solution is prepared by dissolving the cyanide and dextrine separately in part of the water, mixing them, and then gradually stirring in the indium in the form of indium chloride solution. This is done under a

good exhaust hood to take away any cyanide fumes that may be formed. The indium cyanide that at first forms, dissolves in the cyanide upon stirring. The solution is diluted to about half its final volume, allowed to stand for a couple of days, and any solid sediment is filtered off. The dark brown solution is diluted to the full extent and analysed, and the composition adjusted by any necessary addition of sodium cyanide.

The use of distilled water is recommended. The vat should be "non-metallic"—i.e., chemical stoneware, glass, or vitreous enamel or rubber-lined steel. It is useful to rubber cover racks or suspenders except where electrical contact is required. Carbon or platinum anodes are employed.

In operating, an electrolyte temperature of 20/25° C. is used. A very wide cathode current density range is utilised, this being from 10–100 amps/sq. ft., dependent upon the configuration of the work, but employing always sufficient to obtain free gassing at the work surface. Throwing power from the indium cyanide electrolyte is very good, and the deposit formed is very soft and a matt grey in appearance. At the completion of the plating, the work is rinsed in a small drag-out wash from which the indium can be recovered, and then it is thoroughly washed in cold and hot waters, dried off, and lightly polished, if desired.

Assuming tri-volent indium, the cathode efficiency of a new bath is about 60%, and of an old bath about 25%. On the basis of uni-volent indium, which is probably the form in which it is present in the cyanide electrolyte, these values become 20% and 8%, respectively. In other words, the cathode efficiency is of similar order to that in chromium plating.

If the sulphate electrolyte is used for indium plating, it is advocated that work preliminarily plated in cyanide electrolytes, should, after thorough washing, be rinsed in cold 5% sulphuric acid and then thoroughly water-washed before passing to the indium sulphate electrolyte.

The sulphate bath is prepared by direct dissolution of metallic indium in sulphuric acid, a rather slow, but quite simple process. For a litre of solution, 21 gms. of indium as foil is dissolved in a dilute solution of sulphuric acid (15·3 ml. of 95% sulphuric acid in 300 ml. of water). The mixture is heated under reflux for several hours. It is then cooled, filtered, and a solution of 10 gms. anhydrous sodium sulphate is added. The solution is diluted to a litre and the pH value adjusted to 2·0 using dilute sulphuric acid.

The sulphate electrolyte is worked at room temperature. Permissible current density is determined not only by the shape of the work, but also by the nature of the base metal or electroplate undercoat, which appear to influence over-voltage characteristics markedly. Analytical control is rigorously exercised to maintain pH value at 2·0 and the indium content at 2·5–3·0 oz. troy per gallon. The need for chemical control cannot be over-emphasised, because indium anodes are employed, anode efficiency is about 100% and cathode efficiency only 50%. This is counteracted by using some inert anodes (carbon or platinum). The pH check gives a rapid indication of maintenance of correct conditions, because rising metal content would be accompanied by a fall in acidity and, therefore, a rise in pH value.

With the sulphate electrolyte, the soidum sulphate constituent helps to suppress electrolytic dissociation of the indium sulphate, and it also serves to adjust electrical conductivity to a constant level. A voltage drop value between anode and cathode provides a quick check upon conditions, especially after the best electrode spacing and current density have been determined by trial for a specific job.

Regarding analytical control a method for determining the free cyanide in the indium cyanide electrolyte is as follows :—

A 10 ml. sample of the cyanide electrolyte is diluted with 100 ml. of distilled water, and 1 ml. of 10% potassium iodide solution is added. This solution is then titrated with N/10 silver nitrate, the end point being read at the first signs of permanent turbidity. If X ml. of the silver nitrate solution is required, then :—

Free cyanide content of electrolyte = 0·13X oz./American gal.

The indium content is determined on a 1 ml. sample from the electrolyte. The cyanide is destroyed by evaporating with aqua regia. The residual solution is diluted to 100 ml., heated to boiling, and indium is precipitated with ammonia. The precipitate is filtered and washed thoroughly with ammonia water. It is ignited and weighed as indium oxide, In_2O_3. If the weight of the latter is Y grams, then :—

Indium content of the electrolyte = 101Y troy oz./American gal.

The indium content of the sulphate solution can similarly be determined by precipitating the hydroxide. No information is given upon the thickness of indium coating for wear resistance. In any case, it will depend upon the tool and the severity of the job it has to do, at least to some extent, and is best left to be determined by users in relation to their own practice. Commercial coatings in America range from a ten-thousandth of an inch to several ten-thousandths, with the thicker deposits for wear resistance. Thus, the very thin deposits are used over normal cadmium coatings of 3–6 ten-thousandths of an inch for corrosion resistance, being used on iron and steel, and as a base for soft soldering on brass, bronze, copper and the like. Thin coatings of indium are also used over silver plating of thickness from 2 ten-thousandths to 1 thousandth of an inch for reflection.

The advent of indium plating was undoubtedly assisted by the war-time restrictions on nickel and chromium, due to the short supply of the latter. Indium has established its role in producing corrosion-resisting films on practically all the common metals of engineering, ferrous and non-ferrous. It has earned a reputation for its optical properties of reflection, and for its wear resistance. There is nothing to be gained by introducing a new process or finish unless it is superior to that used for the purpose previously, or unless there are economic or production advantages. Some comparisons are warranted on this basis with chromium for tool work. Chromium plating in this field is very much a specialist's job, meticulous care is necessary with pre-cleaning and surface preparation, and high current densities are required for the deposition—viz., from 100 to several thousand amps./sq. ft. Often very thick deposits are necessary, and these require a long plating time, with close temperature control on a warm electrolyte. The final grinding and lapping to size is also intricate, requiring skilled tool makers. Indium coatings can be applied at lower current densities at ordinary temperatures. Throwing power is better and apparently thinner coatings are satisfactory. In any case polishing or

lapping after plating is simple. In the case of corrosion-resisting coatings on iron and steel, chromium has to be applied over very heavy nickel coatings, for which again, special steps have to be taken in preparatory treatments to ensure good adhesion. With indium, very good results are reported over ordinary cadmium coatings. These platings also have the advantage of easy solderability.

Therefore, it is considered that indium plating is a process which merits attention when it becomes available in this country. Undoubtedly, its cost and availability on the basis of the quantity of indium producible will limit the scope for indium plating, but this will also ensure that it is restricted to purposes for which it is pre-eminently suited.

The Chemical Determination of Tantalum and Niobium in Tungsten-Bearing Heat-Resisting Steels

By E. W. Harpham

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Tantalum and niobium (columbium) are being increasingly used as alloying elements in steels and ferrous alloys, chiefly to combat "weld-decay" in welded structures. Methods for determination of these elements are therefore being closely studied, and in this paper a method for their routine determination in tungsten-bearing heat-resisting steels is described.

DURING the past few years, tantalum and niobium (columbium) have found increasing use as alloying elements in steels and ferrous alloys, chiefly as addition agents to combat "weld-decay" in welded structures. Methods for the determination of these elements are therefore becoming more important, and are receiving closer study than in the past.

Methods for the determination of tantalum and niobium in steels and ferro-alloys are now based almost entirely on the classical work of W. R. Schoeller and his collaborators,[1] and their procedures for the separation and determination of these elements are now widely recognised as the most reliable and accurate of those now available.

In any method for the determination of these elements in steels, it must be remembered that both niobium and tantalum are practically always present together. It has been the practice in the past in steel works analysis to obtain the total weight of the mixed tantalum and niobium pentoxides and to calculate either to percentage tantalum or percentage niobium, according as to whether a high-tantalum ferro alloy or a high-niobium ferro alloy has been used as the source of these two elements. That this last assumption can, and usually does, give very unreliable results is clearly illustrated by a study of typical ferro-alloy compositions which are given in Table I.

TABLE I.—ANALYSES OF FERRO-ALLOYS OF TANTALUM AND NIOBIUM.

Sample	1	2	3	4	5	6	7	8
Ta	65·10	64·00	63·10	41·20	15·00	9·17	5·65	3·60
Nb	13·20	13·00	14·80	39·20	59·00	57·68	61·05	55·58
Fe	10·50	12·00	8·80	—	17·50	27·00	—	29·90
C	0·14	0·15	0·11	0·025	0·05	0·10	—	0·31
Mn	5·70	3·50	6·20	—	5·70	—	—	—
Si	0·80	0·60	1·10	—	2·20	0·93	—	5·86
Al	1·60	4·00	1·45	—	2·00	—	—	—
Sn	2·35	2·00	3·70	—	0·25	—	—	—
Cu	0·80	1·00	0·80	—	Nil	—	—	—
W	—	—	—	—	—	3·96	—	4·12
Ti	—	—	—	—	—	0·60	—	0·30

Therefore, unless a chemical separation of the two elements be made, the results as ordinarily calculated will be of very limited value.

Another factor which must be borne in mind is that certain types of heat-resisting steels to which tantalum and niobium have been added may also contain several per cent. of tungsten, and it is a fair criticism to say that most of the published procedures do not take into account the possible presence of this element. The presence also of appreciable amounts of titanium can also cause difficulties.

According to Weiss and Landecker[5] and Lundell and Hoffman[6], both tantalum and niobium are quantitatively precipitated by evaporation with perchloric acid, but it has been pointed out by a number of authors [2-3-6], that quantitative precipitation does not take place in the presence of appreciable amounts ($> 0{\cdot}1\%$) of titanium. The tantalum and niobium bearing steels which have come under the author's notice, and for which the procedure described below was devised, contained only traces of titanium. Silverman[2] describes a procedure for the determination of niobium (and presumably tantalum) by evaporation with perchloric acid. He further states that if the steel is dissolved in dilute hydrochloric acid (1 : 2), a precipitation with cupferron carried out, and the cupferron precipitate treated with nitric acid and fumed with perchloric acid, all of the niobium (and tantalum) will be found in the residue. In the method described by Cunningham[3], the tantalum, niobium and titanium are precipitated by cupferron in reduced solution, the iron removed as sulphide in ammoniacal tartrate solution; tantalum, niobium and titanium again precipitated with cupferron and the precipitate ignited to oxides and weighed. The titanium is determined colorimetrically with hydrogen peroxide and the niobium determined volumetrically by reduction in a Jones reductor in the presence of a known amount of titanium, which latter serves to prevent hydrolysis

1 W. R. Schoeller, "The Analytical Chemistry of Tantalum and Niobium." Chapman & Hall, Ltd., London, 1937.
2 L. Silverman, *Ind. Eng. Chem.*, Anal. Ed., 1934, 6, 287.
3 T. R. Cunningham, *ibid*, 1938, 10, 233.
4 R. Genders and R. Harrison, *Jnl. Iron & Steel Inst.*, 1936, 134 (ii), 173.
5 Weiss and Landecker, *Chem. News*, 1910, 101, 2, 13, 25.
6 G. E. F. Lundell and J. I. Hoffman, "Outlines of Methods of Chemical Analysis." John Wiley & Sons, New York, 1938.
7 W. R. Schoeller and E. F. Waterhouse, *Analyst*, 1924, 49, 215.

of niobium in the reductor, and helps to secure a more complete reduction of the niobium. It is to be noted that in the volumetric determination of niobium, apart from the uncertainty of the degree of reduction of the niobium, the presence of tantalum, except in very small amounts, interferes with the niobium titration, causing low results as shown by Schoeller and Waterhouse.[7] More recently, Schoeller[8] has further criticised volumetric methods for niobium, and considers them no longer justified now that reliable gravimetric methods are available. However, Treadwell and Nieriker[9] have recently shown that 5-valent niobium can be quantitatively reduced to the 3-valent stage using an electrolytic reductor, and describe the oxidimetric determination of niobium by itself and in the presence of iron, vanadium and titanium, but the apparatus is too complicated to warrant its use in a routine steel works laboratory.

Other methods which have been described for the determination of tantalum and niobium include that of Fucke and Daublander[10] who describe the separation and determination of tantalum and niobium in steels by the use of benzene-arsonic acid in sulphuric acid solutions containing hydrogen peroxide. Photometric methods for the determination of tantalum and niobium in steels and ferro-alloys have been described by Thanheiser[11] in which the yellow colour which niobium gives with hydrogen peroxide in a mixture of concentrated sulphuric and phosphoric acids, and the yellow colour given by tantalum with pyrogallol in a 3% solution of ammonium oxalate have been measured photometrically. Klinger and Koch[12] have also described photometric methods for the determination of these elements in steels, and Schliessmann[13, 14] has studied the spectroscopic determination of tantalum and niobium in highly alloyed steels.

The determination of tantalum and niobium in carbon and low alloyed steels not containing tungsten is not a difficult matter, but the presence of appreciable amounts of titanium and tungsten necessitates special and not always straight-forward procedures. The usual procedure for the separation of the earth acids from tungsten as described in two recent books on the chemical analysis of iron and steel[15] is essentially the " magnesia " method of Powell, Schoeller and Jahn[16], but the author prefers to use the " sodium chloride " method exactly as set forth by Schoeller[1], for this separation. This procedure does not require platinum apparatus, no strongly alkaline solutions are in contact with glass beakers, and many observations lead to the conclusion that this method gives a more complete separation of the earth acids from tungsten in one operation than does the " magnesia " process with the amounts of tungsten and earth acids present in these types of steels. By carrying through a series of " blank " determinations using a steel of exactly similar composition but not containing either tantalum or niobium, by both the " magnesia " and " sodium chloride " separation procedures, it was found that the average weight of the blanks obtained in the " magnesia " process was 2–3 mg. and in the " sodium chloride " process the average weight was 0·2–0·5 mg. In the former case the residues consisted mainly of tungstic oxide and silica, while in the latter case no silica or tungstic oxide was found in the residues.

The following method, which is a combination of well-established procedures, has been found to be very suitable for the determination of tantalum and niobium in heat-resisting steels which contain several per cent. of tungsten together with high percentages of chromium and nickel and which come within the following approximate limits :—

	Wt. per cent.
Carbon	Up to 0·5
Silicon	0·5– 2·0
Nickel	12·0–35·0
Chromium	12·0–35·0
Tungsten	2·0– 5·0
Molybdenum	Up to 6·0
Cobalt	„ 15·0
Tantalum and Niobium	„ 3·0

In the steels of this type which have been examined, no zirconium or appreciable amounts of titanium ($< 0·2\%$) have been present, so these elements will not be considered further. Any selenium would be practically entirely eliminated in the initial decomposition of the steel.

Procedure

For steels containing up to 1% tantalum or niobium, use 5 gm. of sample, and for 1–3% use 2 gm. of sample.

Dissolve the specified weight of steel in 50 ml. of concentrated hydrochloric acid with gentle heating. Oxidise the solution with nitric acid and boil off all nitrous fumes. Remove from the hot plate and cautiously add perchloric acid (60%), using 35 ml. for a 5 gm. sample and 25 ml. for a 2 gm. sample. Evaporate the solution to fumes of perchloric acid, and allow to fume gently for 10–15 minutes. Remove the beaker from the hot plate, cool, add 10–15 ml. of hydrochloric acid and warm to dissolve the cake. Dilute to approximately 300 ml. with hot water and allow to stand for about one hour. Filter through a pulp filter, and wash the beaker and precipitate thoroughly with hot 5% hydrochloric acid, using a " policeman " and small pieces of filter paper for cleaning the beaker. Transfer the filter and precipitate to a clean (and preferably new) nickel crucible, and ignite at a temperature just high enough to burn off carbonaceous matter.

Add approximately 2 gm. of A.R. sodium hydroxide pellets to the ignited residue, then a few drops of water, and dry the residue thoroughly on a hot plate. Fuse the mass gently over a small flame until a clear fusion is obtained. A final dull red heat for about one minute is all that is necessary. Dissolve the cold melt in 10–15 ml. of half-saturated sodium chloride solution, warming gently if necessary, and allow the crucible and contents to stand in the cold for several hours or preferably overnight. Filter the contents of the crucible on a small, tight paper pulp pad, and wash with small portions of half-saturated sodium chloride solution until the washings do not react alkaline to litmus. Place the pulp and precipitate in a small 250 ml. beaker, stir to a thin paste with water and add a few drops of hydrochloric acid. Digest on the hot plate just short of boiling until all dark particles have disappeared, adding a few more drops of hydrochloric acid if necessary. Allow the beaker to cool, then filter through a Whatman No. 541, 12·5 cm., paper, washing with hot 10% ammonium chloride solution. Ignite paper and precipitate in a tared silica crucible and weigh as mixed tantalum and niobium pentoxides.

8 W. R. Schoeller, *ibid*, 1942, 67, 321.
9 W. D. Treadwell and R. Nieriker, *Helv. Chim. Acta*, 1942, 25, 474.
10 H. Fucke and J. Daublander, *Tech. Mitt. Krupp, Forschungsber*, 1939, 2, 174.
11 G. Thanheiser, *Mitt. aus dem Kaiser Wilhelm Inst. für Eisenforschung*, 1940, 22, No. 17, 255.
12 P. Klinger and W. Koch, *Archiv. Eisenhüttenwesen*, 1939, 13, 127.
13 O. Schliessmann, *Spectrochim. Acta*, 1939, 1, 239.
14 O. Schliessmann, *Tech. Mitt. Krupp, Forschungsber*, 1939, 2, 185.
15 E. Gregory and W. W. Stevenson, " The Chemical Analysis of Metals and Alloys," 2nd ed., Blackie & Sons, Ltd., London, 1943.
E. C. Pigott, " The Chemical Analysis of Ferrous Materials," Chapman & Hall, Ltd., London, 1941.
16 A. R. Powell, W. R. Schoeller and C. Jahn, *Analyst*, 1935, 60, 506.
17 K. Brüning, K. Meier and H. Wirtz, *Metall und Erz*, 1939, 36, No. 22, 551.

The tantalum and niobium are then separated by the fractional tannin procedure, exactly as described by Schoeller,[1] and will therefore not be repeated. For routine work and for work in which the greatest accuracy is not required, it is sufficient only to retreat the first mixed fraction of tantalum and niobium oxides into a further two fractions; a pure tantalum portion, the remainder being taken as niobium.

The author does not recommend determining tantalum, niobium, tungsten and silicon on the same portion of the sample. Recovery and determination of tungsten involves a further period of 12 hours, and in the initial precipitation of the tungsten, tantalum and niobium, the separation of the tungsten may not always be complete. It is better to determine tungsten in a separate portion of the sample, and to correct the weight of tungstic oxide from the amounts of earth acids found previously.

Summary

A brief review of recently published methods for the chemical determination of tantalum and niobium is given, and then a method for the routine determination of tantalum and niobium in tungsten-bearing heat-resisting steels is described. The importance of determining both tantalum and niobium contents is pointed out.

Acknowledgments

The author wishes to thank Mr. D. A. Oliver, M.Sc., F.Inst.P., Director of Research, Messrs. William Jessop & Sons, Ltd., for his interest in this work, and to various colleagues in the Research Department for their helpful advice.

Reviews of Current Literature

An Introduction to X-Ray Metallography

THE examination of metals, both ferrous and non-ferrous, by means of X-rays is becoming of greater importance to metallurgists generally and is being widely used in works laboratories in the United States. In Britain, too, great progress has been made in the use of X-rays for examining metals, especially in connection with the work of research associations, universities, and in research laboratories of large firms, but, with these and other exceptions, there is still ample room for increased application in the normal routine of manufacture.

X-ray examination of metals has proved itself of great fundamental value in the progress of knowledge of the structure of metals but its importance industrially is only gradually being appreciated in this country, probably due to the fact that there is a dearth of reliable books on the subject. Until quite recently, the technique was in the hands of a few specialists—physicists rather than metallurgists—who lacked metallurgical experience and who were not able to appreciate the metallurgical uses to which the new tool could be put. The stress of war, however, has accentuated the increasing demand for physicists and metallurgists, with a knowledge of X-ray crystallography and for crystallographers skilled in metallurgy, and to-day X-ray analysis is rapidly extending far beyond the original aim of determining atomic arrangements in crystals. It is now proving itself an essential tool in metallurgical research and it is for this highly specialised field of application that this book has been prepared.

It should not be assumed that microscopic analysis, so familiar to the metallurgist, has been relegated to an inferior position with the development of the X-ray method of examination. Both methods have their limitations and should be regarded as complementary to each other. Thus, in the investigation of an equilibrium diagram, X-rays are invaluable in the preliminary stages, but the use of the microscope is necessary to increase the accuracy of the phase boundary determinations. However, the scope of X-ray technique covers quite a wide field of enquiry and is rapidly growing and this book, designed to give the metallurgist an introduction to the application of X-rays, will be invaluable to the research workers in college or in industry and to students preparing for degrees.

The introductory chapter deals with the scope of X-ray analysis, and subsequent chapters are concerned with X-ray generating apparatus; the space lattice; the diffraction of X-rays by the crystal lattice; experimental methods of obtaining diffraction patterns; the influence of the atomic pattern on the intensities of the X-ray reflexions; the crystal structure of the metals; the study of thermal equilibrium diagrams by X-ray methods; the measurement of grain size; grain orientation; application of X-rays to the study of refractory materials; radiography and microradiography; in addition, an appendix deals with the making of Lindermann glass and cellophane capillary tubes for cylindrical Debye-Scherrer powder specimens; gives tables of indices for cubic crystals, for the hexagonal system, for emission, spectra and absorption edges, of X-ray atomic scattering factors, X-ray absorption, and angular factors for various experimental arrangements. The appendix also includes a list of metal compounds, giving their system and structure type, space group, and lattice constants.

It will be noted that the book falls into two main parts, one dealing with the physics of X-rays, and the other with their application to certain aspects of metallurgy. The subject is admirably presented and the author is to be congratulated in combining a guide to the literature with a valuable work of reference. In many respects this book is essentially practical and, besides research workers and students, it will prove a valuable addition to the books of reference needed by the average metallurgist.

By A. Taylor, M.Sc., Ph.D., F.Inst.P., with a Foreword by Sir Lawrence Bragg, O.B.E., F.R.S., Royal 8vo in size, 400 pp., 240 illustrations and 2 loose charts giving standard grain sizes. Published by Chapman and Hall Ltd., 11, Henrietta Street, London, W.C.2. Price 36/- net.

Symposium on Magnetic Particle Testing

PROBLEMS associated with magnetic particle testing are a frequent cause of discussion and this recent symposium on the subject, presented last year at a meeting of the American Society for Testing Materials and embracing eight technical papers, brings to light several aspects which assist in simplifying these problems. The papers dealt with equipment for such testing, specifications for magnetic particle testing and testing procedure, the magnetic particle tests of aircraft materials, of railway materials and of castings and forgings, and miscellaneous applications of magnetic

particle testing. This special volume of the proceedings, in addition to the eight papers and resulting discussion, includes two A.S.T.M. methods of magnetic particle testing and inspection of commercial steel castings and of heavy steel forgings.

In the first paper A. V. de Forest and C. E. Betz discussed equipment for magnetic particle testing and stated that for this type of test the two essentials necessary are a magnetic field of proper direction, character and amount, and the application of the most suitable indicating particles to locate the leakage fields caused by the discontinuities. To meet these requirements, in order to suit applications in various types of inspection, it is necessary to make the proper choice among a number of possible variables. These variables include the direction of the magnetic field (circular or longitudinal), the intensity or distribution of the field (direct current or alternating current), the sensitivity of the method in magnetising and applying the magnetic particles (continuous or residual), the choice of particles, and the application of particles (wet or dry).

Magnetic testing in the field of manufacture of aircraft engines and other parts has favoured the use of the wet method, either residual or continuous magnetisation, and usually direct current, whereas in the overhaul of aircraft and automotive parts and railway equipment it is preferable to use alternating current, residual magnetisation, and the dry method. Inspection of heavy forgings and castings generally calls for the dry continuous direct current method, and the inspection of welds is done by prod magnetisation, using direct or half-wave rectified current and employing portable units. Many other applications require various types of equipment, and in all fields specially built equipment finds considerable use in certain applications.

Magnetic particle inspection, particularly from the standpoint of specification requirements, was dealt with by H. H. Lester. For the detection of surface cracks or similar discontinuities the method is considered one of the best methods available for ferrous material in spite of certain objections. One disadvantage is that it gives undue prominence to surface defects, and another is that it gives occasional trouble in connecting with undercuts at the edges of welds where the pattern of a crack beneath the undercut is masked by the pattern of the undercut and leads to the possibility of incorrect diagnosis. Other disadvantages are associated with procedure. Standards have not yet been set up representing typical patterns of acceptable and unacceptable conditions except for the U.S. Navy, where there are standard illustrations for laminations in rolled plate and other rolled materials. The need for similar illustrations for surface defects in castings and welds is considered necessary.

Lester also considered the application of magnetic testing to the detection of subsurface defects and to the limitation imposed in such testing. In this application of the test, the pattern obtained records variations in the magnetic field due to voids, inclusions or crack-like discontinuities with distinctive characteristics which distinguish them from those due to surface defects. Many variables are involved in the formation of the powder pattern, such as flux density, orientation, depth and dimensions of defect, residual magnetism, magnetic powder characteristics, surface roughness, etc. Because of these variables, establishing pattern characteristics that are distinctive and that show depth below the surface is an extremely difficult problem. Distinguishable patterns must be obtained for cracks, tears, unfused sections, incomplete penetration in welds, laminations in wrought materials, slag or sand inclusions, and voids in castings. To make subsurface detection satisfactory, it is not only necessary to obtain distinctive pattern characteristic for these types of defects, but also to obtain pattern characteristic which will enable an estimate to be made of the depth beneath the surface and the depth dimension of the defect. At the present time satisfactory specification requirements cannot be worked out due to lack of standards.

Specifications and procedures in magnetic testing and the reason for their inclusion in certain A.S.T.M. specifications were also dealt with by A. P. Spooner, who calls attention to the importance of the use of the proper magnetic field direction and strength. Examples are presented of field strength surveys of a few castings and forgings, demonstrating that even on a symmetrically shaped part, significant variations of magnetic field strength occur on the different surfaces. It is also stated that while magnetic particle inspection sometimes replaces older and less searching testing methods, it usually supplements them by by presenting new and previously unrecognised problems. It is a method which demands most careful consideration for proper interpretation, and understanding and co-operation between consumer and producer are primary requisites.

In dealing with the magnetic inspection of aircraft parts, E. O. Dixon discussed briefly the prevailing types of equipment and methods used and also discussed at greater length the grading of steel for the production of aircraft parts with reference to the co-ordination of steel test results and final inspection rejection. Prevailing practice in many aircraft plants is to subject parts to the magnetic test at various stages of fabrication in order to prevent avoidable effort on useless material, and the changing of the amount and type of inspection from time to time because of varying conditions in materials and processes. Methods and standards of inspection should be based on the engineering significance of the part inspected, and the commonly encountered defects to which attention should be given are grinding checks, quenching cracks and non-metallic inclusions. It is also pointed out that variation in magnetisation in regular shapes is not widely recognised, and such variation in some cases results in a wide variation of the sensitivity of the inspection method.

Consideration was given to the application of magnetic inspection to railway materials by L. B. Jones. First used to reveal fractures in locomotive and carriage axles, magnetic testing has progressed rapidly, as the necessary technique was developed for handling more and more complicated shapes, and there are now very few locomotive and carriage parts which are not subjected to such inspection where failures have been experienced. The basic principles of such testing are the same in the railway field as in other industrial application, but testing technique varies and satisfactory results are obtained by a variety of procedures. Some use the wire nail test to determine the proper flux density, and others depend on the appearance of the powder after application as a safeguard against too high flux values. Where longitudinal cracks are suspected, a heavy current is passed, and where transverse cracks are encountered a magnetic flux is generated in an axial direction. This flux may be generated by a few turns of cable with heavy current, or by coils having many turns with small

current. Where "L" or "H" cracks associated with torsion fatigue are encountered, they may be brought out by a magnetic flux generated in outside coils, to ensure uniform flux distribution at the damaged section, and cracks located in bends or corners can be detected by use of a horseshoe or flatiron magnet, which create a local flux through the particular area under suspicion. For all varieties of procedure, wet and dry methods of applying the testing particles or powder are both in general use.

The magnetic inspection of castings and of forgings was dealt with in two papers by C. L. Frear and C. J. Boyle respectively. In the former the utilisation of the method to the detection of such casting defects as sand and slag inclusions, gas porosity, hot tears, cracks, shrinkage, and chaplets and internal chills is considered. Sand or slag inclusions on the surface may be found by visual inspection and those just below the surface may cause sufficient flux leakage to be identified. Cavities from gas porosity near the surface and hot tears which reach the surface are easily identified, while cracks, which are surface defects, are also easily identified by magnetic inspection, although they may be missed entirely by radiography. Massive shrinkage is a deep-seated defect difficult to detect by magnetic inspection except where accompanying tears or cracks approach the surface and, although unfused chills are indicated by magnetic testing, chaplets and chills being usually of lower carbon content than the surrounding metal may give magnetic powder indications even though complete fusion exists.

It is also pointed out that one of the greatest restraints to the general acceptance of magnetic inspection of castings has been the fear on the part of manufacturers that the persons interpreting the magnetic test indications would not be sufficiently experienced to distinguish between harmful defects and indications resulting from incorrect magnetising technique or slight surface defects not affecting serviceability. The Bureau of Ships has limited the use of magnetic powder inspection to the location of surface hot tears and cracks, and in castings where small tears and cracks are permitted, the acceptability of such defects is judged with regard to size, extent and direction by comparison with the Bureau Radiographic Standards. In general, magnetic powder inspection and radiography have been found to supplement each other in the inspection of castings, but neither one by itself shows conclusively the presence of harmful defects in all areas.

In the second paper are described the results obtained from an investigation made to determine some basic information regarding the different methods of magnetisation, field intensity, and sensitivity of powders and pastes in connection with the inspection of large forgings. The investigation was confined to surface defects only, such as forgings fold and laps, thermal cracks that extend to the surface, and inclusions. Several forgings of different sizes known to contain thermal cracks were used for the tests and were thoroughly demagnetised between each series of tests and results obtained were recorded on Scotch tape.

A comparison of the data obtained indicates that circular magnetism is the more effective method, and that circular magnetism with direct current obtained from generators, copper oxide rectifiers, or storage batteries, gives equivalent results. Direct current is more effective than alternating current, and the d-c continuous method is the most sensitive and the most definite for revealing surface defects. The d-c residual method is slightly less effective than the d-c continuous method, while the d-c surge method is between the d-c continuous and the d-c residual method in effectiveness. Electromagnets may be used in some cases, but they are not as effective or efficient as circular magnetisation. The wet method is more sensitive than the dry method.

In an 18 month period, magnetic tests were made on 3092 forgings, ranging in weight from a few hundred pounds to 75 tons. Of this number, 130 or 4·2% were rejected for defects revealed by the test. Unfortunately, most of the defects so located could not be detected until after final machining, and this limitation in testing resulted in a loss of man-hours and machining operations that were, in most cases, of greater value than the forging itself.

Miscellaneous applications of the magnetic particle test were discussed by E. A. Snader in the last paper presented at the Symposium. One of the first and most important applications of the test was in testing turbine blading. In the testing of such parts, as a final inspection before installation, the wet method is used, but for the detection of fatigue cracks in blades tested after a period of operation in the turbine the dry method is preferred. For the detection in gear milling hobs and milling cutters of certain types of defects which, partly due to complex design, are difficult to detect, the fluorescent particle method is preferred. In testing forged pinion teeth, the magnetic field must be properly applied. Either small electromagnets or a strong circumferential field, obtained by passing sufficient current directly through the pinion, give very satisfactory results. A circular current is also used for testing bolts and studs for cracks in the roots of threads and here also the fluorescent method can be used with advantage. Magnetic testing is also of use for determining laminations in the edges of heavy steel plates, fire cracks and stress corrosion cracks in boiler plates, defects in tools, and for inspecting welds for soundness.

Copies of this 130-page book, 6 in. × 9 in. page size, heavy paper cover, can be obtained from A.S.T.M. Headquarters, 260, South Broad Street, Philadelphia 2, Pa., at $1·25 each.

Uses of Briquetted Alloys

"IMPROVING Cast Iron with 'St. Lawrence' Briquetted Alloys" is a booklet recently published by St. Lawrence Alloys & Metals, Ltd., Beauharnois, Quebec, Canada. This booklet, one of a series of technical publications available from the Company, tells how foundrymen can produce good castings, in spite of variations in raw materials, by adding briquetted alloys of silicon, manganese, or chromium to the cupola. The metallurgical and physical properties of good cast irons are discussed.

One section of this booklet gives recommended compositions for agricultural, mining, railroad, and machinery castings, and stove and furnace parts. Tables show how to calculate precisely cupola charges and alloy additions in both metric and English units. Photomicrographs and charts are used to show the advantages of briquet additions.

Copies of this 20-page illustrated booklet, now available in English, Spanish, Portuguese, French, and German editions, will be sent to foundrymen if they will write on their business letterheads. Interested readers are requested to specify the language most useful to them.

Correspondence

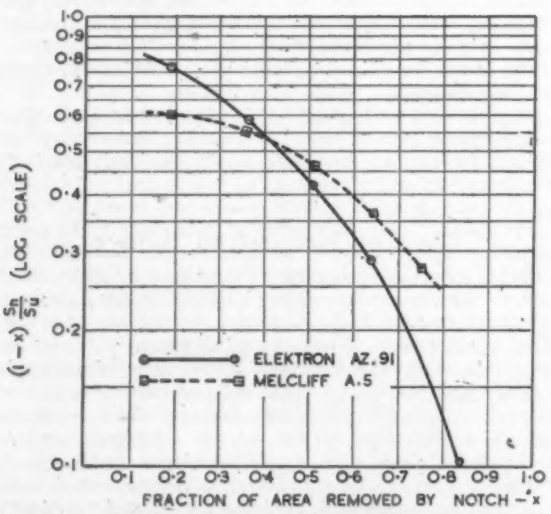
A Discussion on the Notch Impact Test and its Interpretations*

To the Editor, METALLURGIA.

Sir,

Mr. Fisher's statement, "Notch sensitivity of a given material varies with the size and shape of a notch" emphasises clearly that the term, "notch sensitivity" requires more suitable definition. If a material does possess a property which may be characterised as "notch sensitivity" we feel that the definition of the property should not vary with the size and shape of the notch. To promulgate a suitable definition, the following points are made :—

The term "notch sensitivity" is expected to relate the change in a mechanical property by virtue of the notch being present. If the separate mechanical properties of : (*a*) ultimate tension strength ; (*b*) 0·2% offset yield strength ; and (*c*) elongation of a 2-in. gauge length are examined for any material with any specific notch, it will be observed that the experimentally determined yield strength has increased, the 2-in. gauge length elongation has decreased, and the ultimate tension strength has increased or decreased (depending on the material and the notch geometry) in comparison with the properties of the unnotched material. It is thus evident at once that since the term "notch sensitivity" relates the change in a mechanical property it will be necessary to stipulate which property is being discussed when the effect of the notch is being discussed. To speak of "static stress notch sensitivity" or "static strain notch sensitivity" would then more clearly define the area of discussion.



A replot of data on other co-ordinates

As Mr. Fisher's Figs. 6 and 7 indicate, "static stress notch sensitivity" of Elektron AZ. 91 and Melcliff A. 5 compare differently with each notch depth and a forthright statement of which material is the more notch sensitive seems impossible to state. A replot of these data on other co-ordinates, however, indicates that a direct comparison for all notch depths can be made as per the accompanying figure.

Notch depth is plotted as the abscissa, but notch depth is expressed as the fraction of the original cross-sectional area which the notch removed :—

$$x = 1 - \frac{(D-2d)^2}{D^2}$$

with D = bar diameter.
d = notch depth.

Notched strength is plotted as the ordinate, but notched strength is expressed as the ratio of the gross area notched strength to the unnotched ultimate strength :—

with Sn = notched area notched strength.
x = fraction of original area removed by notch.
Su = ultimate strength in unnotched condition.
$(1-x)\ Sn$ = gross area notched strength.

In this co-ordinate system the "index of static stress notch sensitivity" could be taken as the slope of the curve and the intercept on the ordinate. Melcliff A. 5 would be classified as the more "stress notch sensitive" of the two alloys by virtue of its intercept on the ordinate.

In this system of comparing static strength specimens of notched aluminium alloys, the plotted curve is approximately a straight line passing through the origin. For such cases the slope of the line furnishes a "stress notch sensitivity index" that lends itself readily to a comparison of materials. This manner of treating experimental data has been discussed by the writer in a paper submitted to the Institute of Aeronautical Sciences (U.S.A.) for early publication.

Jan. 29, 1946. Sincerely yours,

Douglas Aircraft Co., Inc., Leo Schapiro,
Santa Monica, Chief Metallurgist
California. Douglas Research Laboratories.

Notch Sensitivity

To the Editor, METALLURGIA.

Dear Sir,

Engineers and metallurgists will feel generally sympathetic towards Mr. Schapiro's desire for an expression which for a given material would always give the same value—or reasonably the same value—for the notch sensitivity "index," irrespective of the size or shape of the notch. On asking a simple and definite question as to whether the static tensile strength of a material is sensitive to the presence of a notch or not sensitive to the presence of a notch, a practical man hates to be told that it all depends. However, it proves to be neither the definition nor the expression that tells him this, but the material itself!

Thus, it is just a simple fact that whereas Melclif A.5 notched bars tested in static tension show a tensile strength (on the reduced area) which first falls below and then rises above the original unnotched value as the notch depth is increased, Elektron Az. 91 notched bars show a tensile strength which falls continuously. It does not seem that any re-arrangement of formulae could alter this fact.

Mr. Schapiro realises this, of course, but rightly comments that there are other properties besides tensile strength which are affected by the presence of the notch: he also points out that to define notch sensitivity adequately and correctly it will be essential to mention which particular property is referred to when a value for notch sensitivity is given.

* See *Metallurgia*, September and October, 1945, pp. 192-198 and 281-284.

These facts are undoubtedly relevant and correct, but nevertheless it does not seem that by choosing either (*a*) some particular property on which the notch sensitivity does not vary differently for different materials, or (*b*) some particular manipulation of dimensions resulting in an approximately invariable notch sensitivity index, that one would be justified in offering that special value as a typical index of notch sensitivity (apart altogether from the impossibility of making such a choice) for the very reasons mentioned by Mr. Schapiro.

Investigations on notch sensitivity up-to-date have naturally been largely concerned with the effects of the notch on the property of tensile strength, and it was this effect which was dealt with in the comparison of Melclif A.5 and Elektron AZ.91 in the original article. On the result of that comparison (and other similar comparisons for different materials) we are driven either to accept a varying notch sensitivity index (for tensile strength) according to the size of the notch, or to compare hopefully the effect of the notch on some other tensile property (or property relationship) than the tensile strength on the reduced area. If the latter method is chosen, what value is to be given for the notch-sensitivity index relevant to the tensile strength? It does not seem to the writer that the difficulty will have been overcome, or even avoided.

A detailed examination of the method suggested by Mr. Schapiro seems to show that it would not by any means achieve the desired result. If the two curves he gives are completed up to the "y" ordinate, it will be found that they both intercept the ordinate at the same point, $y = 1{\cdot}0$, as all such curves would, therefore the intercept cannot be used as an index. Also, the complete curve for Melclif A.5 changes its gradient from about 3 : 1 at $y = 1{\cdot}0$, to about 1 : 3 at $y = 0{\cdot}6$, and back again to 1 : 1, at $y = 0{\cdot}8$. Thus, if slope were chosen for use as index, it would vary considerably with the size of the notch.

In general, the writer feels that any interpretation of mechanical property phenomena should tend towards rationalisation and preferably simplification, and that the result of any formula manipulation or mathematical expression of property relationships should be interpretable or explainable on a rational basis. It is on such a basis that the writer's formula for notch sensitivity index (for tensile strength) is derived.

Yours faithfully,

Magnesium Elektron, Ltd., ARTHUR FISHER.
Clifton Junction.
Feb. 14, 1946.

The Elastic Modulus

Single Crystals v. Polycrystalline Aggregates

The Editor, METALLURGIA.

Sir,

There have recently been published some extremely interesting data on the elastic anisotropy of highly-oriented magnetic strip steel.[1] Material of this type is characterised by an exceptionally high degree of preferred orientation of [100] crystal directions along the direction of rolling, the general nature of the crystallographic arrangement being one in which the close parallelism of [100] crystal directions with the rolling direction is associated with [110] planes virtually parallel to the rolling plane, thus yielding a preferred orientation of [110] crystal directions transverse to the rolling direction, coupled with a preferred orientation of [111] crystal directions at angles of approximately 55° from the rolling direction.

The data referred to are summarised in the Table below: they show that the directionality of the elastic behaviour of polycrystalline aggregates of the type described is indistinguishable from the directionality of the elastic behaviour of single crystals of pure iron.

ANISOTROPY OF ELASTIC MODULUS IN MAGNETIC STRIP STEEL
3·2%—SILICON, 0·025 IN. THICK
PULLING SPEED, 0·025 IN. PER MIN.

Angular Deviation from Rolling Direction	0°	90°	55°
Crystallographic Direction	[100]	[110]	[111]
Young's Modulus, lbs./sq. in. $\times 10^{-6}$ ..	18/19	30	38·6/41·0

Arguing from the simplest possible version of the domain theory of ferromagnetism, I have shown elsewhere[2] that by assuming that the externally-measured magnetic behaviour of a polycrystalline aggregate (of iron or an iron-silicon alloy) will reflect the internal magnetic behaviour of three hypothetical "crystal-groups" which, respectively, simulate [100], [110], and [111] characteristics, it seems possible to devise a direct and easy "trichotomy" of the technical magnetisation curve which leads to quantitative estimates of the group-distribution of crystal orientation in close agreement with those arrived at by more familiar but less simple methods.

By dissecting the magnetisation curve in this way, I have found that for hot-rolled sheet steel, which is usually regarded as a random aggregate, the actual quantitative distribution of crystal orientation over the hypothetical [100], [110], and [111] crystal groups consistently simulates the approximate order of 2 : 2 : 1 (see Table II, I.S.I., 1944, *loc. cit.*).

Adopting the view that the externally-measured elastic modulus of a hot-rolled sheet aggregate should similarly tend to reflect the joint elastic behaviour of the same three hypothetical crystal groups, the means values of the "internal" moduli shown in the Table would lead us to expect an externally-measured modulus of approximately

$$\left[\{(18{\cdot}5 \times 2) + (30 \times 2) + (39{\cdot}8 \times 1)\}/5\right] \times 10^6 = 27{\cdot}4 \times 10^6 \text{ lbs./sq. in.}$$

And this is precisely the order of things that we consistently observe in practice on hot-rolled sheet aggregates. For truly random orientation, defined as connoting the equivolumetric distribution of the three hypothetical crystal groups, we should expect a higher value of the more familiar text-book order of $29{\cdot}4 \times 10^6$ lbs./sq. in.

There is nothing new in the idea of evaluating the elastic modulus of a polycrystalline aggregate by the integration of individual crystal behaviour. I cannot trace, however, that it has previously been suggested that in polycrystalline aggregates of material which exhibits marked elastic anisotropy in single-crystal form, there may exist a measure of "group-behaviour" which will render simple measurements of the elastic modulus competent to furnish a supplementary or even a substitute method of assessing the general nature of the build-up of the operative orientation along different directions.

I am therefore submitting these few observations in advance of further experimental work on ferromagnetic material because the possibilities would not appear to be unattractive in fields concerned with those even

1 Benford, "Metal Progress" (American Society for Metals), 1946, January p. 94.

2 Richer, *Journal of the Iron and Steel Institute*, 1944. (Advance Copy, September, 1944).

more "sensitive" lattices, such as copper, brass, etc., where the elastic antisotropy is appreciably greater than in iron and steel, and in which, in the absence of ferromagnetic phenomena, the reliable assessment of directional differences in the build-up of the operative orientation demands recourse to the laborious technique of X-ray pole diagrams.

Orb Iron Works, Newport, Mon. Feb. 28, 1946.

Yours faithfully,
G. C. RICHER.

Some Recent Appointments

Mr. J. B. THOMAS, F.C.A., who became General Manager of Hadfields, Ltd., on the retirement of Sir Peter Brown and Major A. B. H. Clerke, has now been appointed Managing Director of the company. Mr. Thomas became associated with Hadfields in 1917, and from 1920 was Confidential Advisor to the late Sir Robert Hadfield.

Mr. W. KILPATRICK has been appointed Managing-Director of J. and H. McLaren, Ltd., and of Kitson and Co., Ltd. He was formerly Managing Director of Davy and United Engineering Co., Ltd.

Mr. LEWIS CHAPMAN, Managing Director of William Jessop and Sons, Ltd., has succeeded Mr. C. W. Kayser as Chairman of the Sheffield Stainless Steel Manufacturers' Association. Mr. Kayser has resigned owing to ill-health.

Mr. F. WOODFIELD has been appointed Works Manager of the Park Gate Iron and Steel Co., Ltd. Mr. J. W. Carr, Mr. D. E. Nutbrown, and Mr. W. S. Allen have been appointed Commercial Manager, Secretary and Chief Engineer, respectively.

Mr. GEORGE BAKER has succeeded Mr. John W. Baker as Chairman of John Baker and Bessemer, Ltd. Mr. Edward Baker and Mr. S. E. Baker have been appointed Joint Managing Directors.

Mr. STANLEY EVANS has accepted an appointment as Secretary of Associated British Engineering, Ltd., and its subsidiaries. Mr. Evans was, for 16 years, Secretary of Darwins, Ltd.

MAJOR STEWART OWEN, M.B.E., who has recently returned from the Forces, has been appointed Acting Area Manager for Scotland, for David Brown and Sons (Huddersfield), Ltd. The announcement has been made as a result of the death, at his home in Glasgow, of Mr. John Alexander Brown, Scotland Area Manager for this firm.

Dr. E. VOCE, M.Sc., Ph.D., has been appointed Metallurgist to the Copper Development Association.

A native of Birmingham, Dr. Voce entered the General Science Course of the City's University in 1920, and graduated B.Sc. in the Honours School of Chemistry three years later. He devoted three further years to metallurgical research in the University, mainly on copper-silicon and copper-silicon-manganese alloys, and received the degrees of M.Sc. and Ph.D.

For nearly four years he was in charge of the casting shops of an important Birmingham firm specialising in the production of cupro-nickel, nickel silver and brass. In 1930, he joined the staff of the British Non-Ferrous Metals Research Association. All the major investigations undertaken by Dr. Voce have been concerned with copper and its alloys, dealing with such subjects as copper base alloys for service at elevated temperatures, the prolonged annealing of hard-drawn high-conductivity copper wires at tropical temperatures, the mechanism of the embrittlement of copper by bismuth, and the stress corrosion and cracking of brasses in sea water, ammoniacal atmospheres and molten metals.

During the war, he carried out a long series of investigations on copper for shell bands on behalf of the Ministry of Supply, and as metallographist to the B.N.F.M.R.A., has made countless examinations of failures and miscellaneous problems submitted by members.

Mr. J. WHITEHEAD has been appointed General Manager of David Brown Tractors, Ltd., with a seat on the Board.

Mr. Whitehead has had considerable experience of large business concerns at home and abroad and has been with the David Brown & Sons (Huddersfield), Ltd., organisation a number of years, during which time he has taken part in the replanning of several of the larger firms in the Group, including the well-known Penistone foundry.

Mr. F. B. MARSH, commercial manager of David Brown Tractors, Ltd., has been appointed a Director of that firm. Joining the David Brown Group in 1936, Mr. Marsh was actively associated with gearing and power transmission development at Park Works, Huddersfield.

Rolls-Royce Factory, Hillington

THE future of the Rolls-Royce factory, at Hillington, has been under most careful consideration by the Government. During the war, the main Hillington factory was devoted to the production of aero engines. The demand for these has, of course, declined enormously, and is now only a small fraction of what it was at the peak war period. It would be highly uneconomic to divide the present smaller volume of work between two factories, and as the Rolls-Royce Company's main works has always been at Derby, and as their research and development organisation is centred there, the company are naturally anxious to concentrate the remaining volume of new construction at Derby.

In these circumstances, the Government have come to the conclusion that it would be in the best interests of Scottish industry and employment to allocate the Hillington factory to other production industries. Arrangements have therefore been made between the Board of Trade and the Ministry of Supply and Aircraft Production to hand over the whole of this factory, with the exception of the foundry, which has been acquired by Renfrew Foundries, Ltd., to Scottish Industrial Estates, Ltd., the non-profit-making Government-financed company which operates the industrial estates in Scotland.

The three blocks already declared surplus to Rolls-Royce requirements are being taken up by the Hoffmann Manufacturing Company, Ltd., for the manufacture of ball-bearings; Scottish Co-operative Wholesale Society for the production of boots and shoes and Atholl steel houses. In the meantime, the Rolls-Royce Company will remain in occupation of those parts of the factory still required by them; but arrangements are being made to ensure the progressive release of further blocks for re-allocation at the earliest possible moment to other industries and firms. Two blocks will become available during the summer, and applications for them are now being considered.

A Metallurgical Study of German Aircraft Engine and Airframe Parts

This report constitutes a summary of further data, resulting from the metallurgical examination of German aircraft engine and airframe parts by the Aero-Components Sub-Committee of the Technical Advisory Committee of the Special Alloy Steel Committee formed for that purpose.

The types examined represent a comprehensive range of various types of German aircraft which have fallen into the hands of the R.A.F. from 1942 onwards.

The principal object of these investigations was to obtain data on the types and quality of materials used, methods of manufacture, efficiency of the heat-treatment to which the parts have been submitted, together with any other information which might prove of value, as, for example, details of the finish. Further, the influence of restrictions, due to our blockade, on German procedure and selection of materials was kept in mind. Attention was given chiefly to engine parts but a number of airframe and miscellaneous components were included. Special features concerning design had been noted in certain instances, but these were not the primary object of the investigations. The Sub-Committee responsible for these investigations and for this report comprise ; Mr. S. Barraclough, United Steel Companies, Ltd. ; Mr. H. Bull, Messrs. Brown-Bayley's Steelworks, Ltd. ; Mr. H. H. Burton, the English Steel Corporation, Ltd. ; Mr. W. J. Dawson, (Chairman), Hadfields, Ltd. ; Dr. R. Genders, M.B.E., S.T.A.M., Ministry of Supply ; Mr. H. J. Hipkins, Royal Aircraft Establishment : Mr. N. H. Mason, R.A.E., Farnborough; Mr. D. A. Oliver, Messrs. Wm. Jessop and Sons, Ltd. ; Mr. L. Rotherham, Thos. Firth and John Brown, Ltd. ; Dr. H. Sutton, Ministry of Aircraft Production ; Miss M. K. Walshaw, Secretary, Brown-Firth Research Laboratories.

The work included in this report embraces the results of investigations carried out as a continuation of the data already published. It has been carried out meticulously. Naturally in this Report no comparisons are made with corresponding parts in British and American Aircraft, neither are certain aspects which our investigations have shown to be open to criticism emphasised.*

Section XIII—Airframe Components

Continued

C. Parts of Undercarriages

Undercarriage struts and tailwheel parts from the following aircraft have been examined :

1. Focke Wulf 190 (undercarriage strut, Item 173 ; tailwheel fork, Item 174).
2. Messerschmitt 210 (undercarriage strut, Item 273 ; tailwheel unit, Item 275).
3. Dornier 217E (undercarriage strut, Item 283 ; tailwheel hub, Item 284).

The chemical compositions and other essential details of the main components of the undercarriage struts are given in Table II. Similar details of the tailwheel parts are given in Table III.

1. *F.W. 190 Undercarriage Strut.*

The damaged assembly is shown in Fig. 22, and the component parts of the strut are shown after dismantling in Fig. 23. The strut is suspended by an end pivot, attached at an angle of 110° to the top of the cylinder. The end pivot is mounted in two bearings in the wing, the front one being in the main spar, and the rear one in a bridge member between two ribs.

As shown in Table II, nearly all the main parts had been made from basic electric-arc steel containing 0·25–0·28% carbon, and approximately 1% chromium and ¼% molybdenum ; they had been used in the hardened and tempered condition, and gave hardness values in the range 300–400. The plunger tube had been coated with 0·0045 in. of chromium, and was the only part to contain vanadium (0·20%), molybdenum being omitted. As in the Messerschmitt 110 strut previously examined, the knee piece and axle had been made as a single hollow casting. A tensile test on a specimen from this part gave values 79 tons/sq. in. for maximum stress,

Fig. 22.—F.W. 19 undercarriage strut.
Fig. 23.—Dismantled F.W. 190 undercarriage strut.

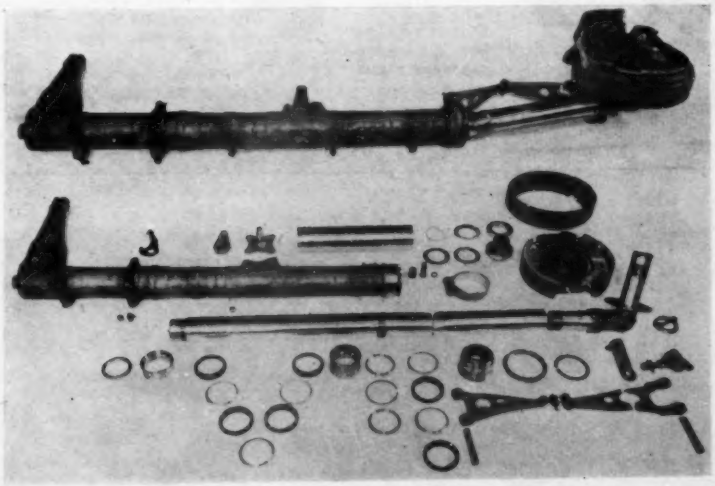


TABLE II—CHEMICAL COMPOSITIONS, ETC., OF MAIN UNDERCARRIAGE COMPONENTS

Item No.	Component	Composition, %										Grain Size	Diamond Pyramid Hardness	Remarks
		C	Si	Mn	S	P	Ni	Cr	Mo	V	Cu			
173	Focke Wulf 190 Undercarriage Strut :—													
	Plunger tube	0·26	0·35	1·06	0·005	0·011	0·05	0·91	Nil	0·20	0·06	3–4	403	Chromium plated, 0·0045 in. thick
	Cylinder	0·26	0·31	0·56	0·005	0·016	0·29	1·14	0·22	Nil	0·15	3–6	384	Hollow forging
	End pivot	0·25	0·25	0·53	0·010	0·009	0·09	1·04	0·14	Nil	0·15	2–4	310–321	Machined from bar. Flash-butt welded to cylinder
	Collar bearing :—collar	0·28	—	—	—	—	0·14	2·52	0·06	—	—	—	388–390	
	shoes	0·27	—	—	—	—	0·04	1·02	0·22	—	—	2–6	382–390	
	Knee piece and axle	0·28	0·47	0·57	0·008	0·012	0·07	0·95	0·19	Nil	—	4–6	351–382	Single hollow casting
	Torsion hinge	0·40	0·25	0·63	0·021	0·008	0·28	0·85	0·24	Nil	0·15	2–4	342–360	Forging
	Retraction bracket	0·25	0·22	0·65	0·015	0·009	0·12	1·00	0·21	Nil	0·13	5–7	297–369	Upset
	Brake drum	0·70	0·23	0·86	0·029	0·028	Nil	0·03	Nil	Nil	0·04	—	272–302	Hot worked and machined
273	Messerschmitt 210 Undercarriage Strut:													
	Plunger tube	0·29	0·34	0·58	—	—	0·07	2·64	Tr.	0·18	—	1–4	381	Chromium plated, 0·0038 in. thick
	Cylinder	0·31	0·34	1·06	—	—	0·06	0·83	Tr.	0·18	—	3–6	286–307	Forging
	Knee piece and axle	0·25	0·28	0·51	—	—	0·08	1·10	0·26	0·01	—	2–5	264–283	Forging
	Torsion hinge	0·28	0·28	1·12	—	—	0·07	0·81	—	0·20	—	3–6	351–400	Forging
	Retraction arm and collars	0·25	0·28	0·56	—	—	0·08	1·00	0·24	Tr.	—	4–7	287–313	Welded preformed sheets
	Radial supports :—cylinder	0·27	0·36	0·53	—	—	0·12	1·02	0·25	Tr.	—	4–6	287–296	Arm welded to cylinder
	arm	0·26	0·17	0·58	—	—	0·09	1·01	0·29	Tr.	—	4–6	331	
	Universal joint housing	0·25	0·33	1·03	—	—	0·07	0·77	—	0·19	—	1, 5–7	304	Casting
283	Dornier 217E Undercarriage Strut :—													
	Plunger tube	0·30	0·34	0·62	—	—	0·09	2·61	Tr.	0·27	—	—	366	Chromium plated, 0·002 in. thick
	Cylinder	0·28	0·28	1·05	—	—	0·08	0·74	Tr.	0·13	—	—	325	Forging
	Threaded shank	0·29	0·31	1·03	—	—	0·02	0·74	Nil	0·16	—	4–7	301–318	
	Locking rings	0·37	0·21	0·54	—	—	0·04	1·54	0·22	0·19	—	—	318–321	
	Axle housing	0·25	0·30	1·02	—	—	0·04	0·79	Nil	0·16	—	—	220	Forging

and 17% for elongation. The end pivot had been machined from bar and flash-butt welded to the cylinder without subsequent heat-treatment.

In general, the steels were slightly dirty and inferior in cleanness to the usual German standard. The molybdenum content was somewhat lower than that of the steels used for similar parts examined earlier.

2. *Me* 210 *Undercarriage Strut.*

A general view of the strut is given in Fig. 24, the component parts are shown dismantled in Fig. 25 (with plunger tube attached to knee piece and axle). The general internal assembly was similar to that of the earlier Me 110 strut, but the design of the outer retraction and supporting members was different.

The plunger tube, machined from a solid bar, had been made from 2¼% chromium-vanadium steel, but the other steels used were of two types (Table II): (*i*) containing approximately 0·30% carbon, 1% manganese, ¾% chromium and 0·20% vanadium; (*ii*) containing 0·25% carbon, 1% chromium and ¼% molybdenum. Almost all the steel parts were used in the hardened and tempered condition, and had a diamond hardness in the range 260–400.

The axle and knee piece had been machined from a single solid forging. A flange was welded round the axle, the whole being hardened and tempered after welding. The retraction arm, fabricated by welding preformed sheets, and the radial support, consisting of an arm welded to a short cylinder, had also been heat-treated after welding.

Fig. 24.—Me. 210 undercarriage strut.

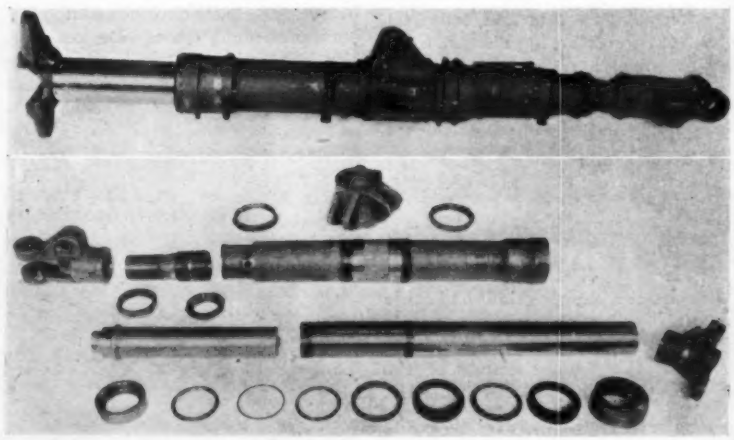


Fig. 25.—Dismantled Me. 210 undercarriage strut.

Fig. 26.—Do. 217E undercarriage strut.

Fig. 27.—Dismantled Do. 217E undercarriage strut.

Fig. 28.—F. W. 190 tail wheel fork.

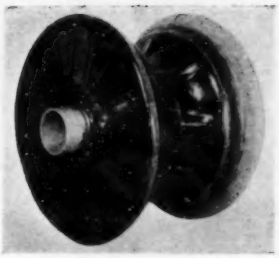


Fig. 29.—Me. 210 tail wheel hub.

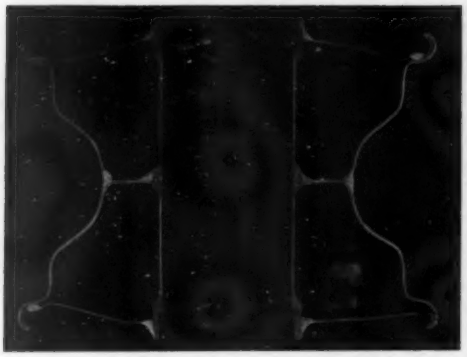


Fig. 30.—Section through Me. 210 tail wheel hub.

Tensile tests on specimens cut from the main components of the strut gave the following results:

	Max. Stress Tons/sq. in.	Elongation %
Plunger tube	78·0	21
Cylinder	60·5	22
Knee-piece and axle	57·5	24

The quality of the steel and standard of workmanship of the various parts of this strut were maintained at a high level.

3. *Do* 217 *E Undercarriage Strut.*

In the Dornier 217 aircraft, each of the main wheels is carried by two struts, braced by tubular members, and retracting backwards into the engine nacelle. One of the struts is shown assembled in Fig. 26, and dismantled in Fig. 27.

The main cylinder, threaded shank, and axle housing were made from a steel containing 0·30% carbon, 1% manganese, ¾% chromium, and 0·15% vanadium. The plunger tube had been made from a 2¼% chromium-vanadium steel, and the locking rings from 1¼% chromium - molybdenum - vanadium steel. The axle housing was in the normalised condition, and had a diamond hardness of 220. The other parts had been hardened and tempered, and possessed hardnesses in the range 301–366. The end pivot and retraction bracket had been made from light alloy of the duralumin type.

There was no deterioration in the quality of the materials used in this component, but an effort had been made to conserve molybdenum.

4. *F.W.* 190 *Tailwheel Fork.*

The fork is shown in Fig. 28. Each leg was a taper flattened tube made in two halves from formed sheet, joined by continuous welds in the centre of the inner and outer faces. The two tubular limbs were welded to one another, and to a hollow stem, and a superimposed strap was welded round the fork where it joined the stem. Each of the two open ends of the fork had a ring welded to it, forming a bearing for the tailwheel spindle.

Reference to Table III shows that the sheets and hollow stem had been made from a steel containing 0·30% carbon, 1% manganese, and 0·6–0·8% chromium, whereas 1% chromium-molybdenum steel had been used for the rings. Stem and rings had been machined from parallel stock.

The diamond hardness of the sheet was 405–425, and of the welds 340–380. These high values, the absence of the characteristic as-rolled structure and hardness gradient, and the uniformly scaled condition of the inside of the tubes, indicate that the fork had been hardened and tempered as a complete unit after welding. The original hardness of the sheets and welds were reproduced in the laboratory by tempering at 400–450° C. after hardening.

5. *Me* 210 *Tail Wheel Unit.*

The hub (Fig. 29) consisted of a

TABLE III—CHEMICAL COMPOSITIONS, ETC., OF TAIL WHEEL COMPONENTS.

Item No.	Component	Composition, %										Grain Size	Diamond Pyramid Hardness	Remarks
		C	Si	Mn	S	P	Ni	Cr	Mo	V	Cu			
174	Focke Wulf 190 Tail Wheel Fork:—													
	Sheet	0·27	0·26	1·08	0·012	0·018	0·07	0·68	Tr.	—	—	5	405–425	Fork heat treated after welding
	Weld Metal	0·20	0·18	0·65	0·011	0·017	0·11	0·69	0·11	—	—	—	340–380	
	Hollow stem	0·33	0·39	1·04	0·012	0·024	0·07	0·84	Tr.	—	—	6	383–414	Machined from tube
	Fork end bearing ring	0·29	—	0·49	—	—	0·14	1·13	0·26	—	—	4	—	
275	Messerschmitt 210 Tail Wheel Unit:—													
	Hub—sheet	0·14	0·24	1·34	0·008	0·008	0·09	0·05	0·06	—	0·27	—	161–236	Pressed steel sheets welded to central tube: no subsequent heat-treatment.
	central tube	0·30	0·22	1·01	0·025	0·019	0·04	0·66	0·05	0·12	0·06	—	221–230	
	Fork—tube (leg)	0·21	0·30	1·09	0·009	0·018	0·07	0·83	0·01	0·10	0·13	—	331–344	Heat-treated after welding
	lug	0·23	0·23	0·55	0·011	0·011	0·02	1·05	0·17	—	0·03	—	278–305	
	Frame—vertical tube	0·25	0·34	1·26	0·010	0·020	0·12	0·78	0·01	0·14	0·10	—	339–360	Heat-treated after welding
	lug	0·24	0·22	1·16	0·018	0·021	0·04	0·68	0·01	0·08	0·10	—	—	
	Ring spring (outer)	1·00	0·82	0·93	0·027	0·016	0·14	0·48	0·15	—	0·12	—	460–511	
	Main tube	0·25	0·22	0·64	0·012	0·018	0·08	1·06	0·13	—	0·06	—	343–368	
284	Dornier 217 Tail Wheel Hub:—													
	Body sheet	0·09	0·23	1·43	0·010	0·012	0·01	0·05	0·03	—	0·17	—	133–152	Formed sheets welded to central tube. No heat-treatment after welding
	Corrugated end sheet	—	0·21	1·43	—	—	0·09	0·08	0·16	—	0·25	—	161–209	
	Central tube	0·33	0·24	0·51	0·019	0·009	0·05	0·07	0·03	—	0·16	—	240–254	
	Bearing housing	0·30	0·22	0·98	0·016	0·019	0·12	0·78	0·03	0·10	0·16	—	296–302	Welded to central tube.
	Outer ball race	0·58	0·25	0·37	0·012	0·019	0·02	1·40	0·01	—	0·13	—	810–823	
	Ball	—	0·27	0·22	—	—	Nil	0·64	0·01	—	0·21	—	902–907	

hollow drum built up from pressed sheets of low carbon 1·3% manganese steel in the as-rolled condition, assembled by welding to a central tube. The hub had not been heat-treated after welding.

The fork had been made in a similar manner to the F.W. 190 fork—*i.e.*, from formed sheets of 1% manganese, 0·8% chromium, 0·1% vanadium steel welded together. This component had been hardened and tempered after welding.

Fig. 31.—Do. 217 tail wheel hub.

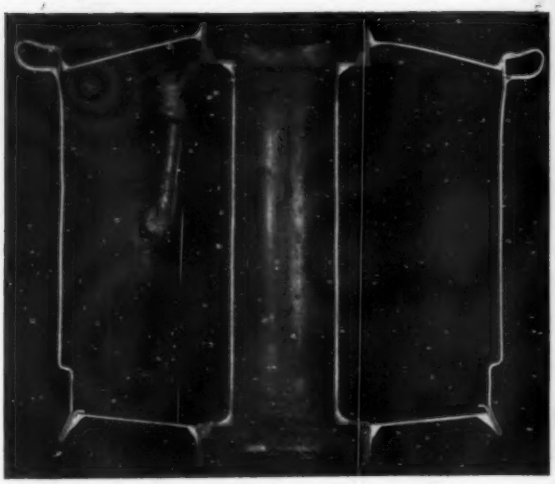


Fig. 32.—Section through Do. 217 tail wheel hub.

Two sizes of ring spring were used in the shock absorber; these had the same composition, 1% carbon, 0·8% silicon, 1% manganese, 0·5% chromium, and 0·07–0·18% molybdenum. The inner rings (570–636 D.P.H.) were harder than the outer rings (460–517).

6. *Do.* 217 *Tail Wheel Hub.*

This component, shown in Fig. 31, consisted of a hollow drum built up by welding together a body sheet and two corrugated end sheets (1½% manganese steel), and a central tube (0·33% carbon steel). A steel ring forming a bearing housing had been welded to each end of the central tube. A section through the hub showing the various joints, is shown in Fig. 32. Details of the analyses of the various parts are given in Table III.

No heat-treatment had been carried out after welding.

D. Parts of Control Systems

The following parts from a Messerschmitt 210 aircraft have been examined:

1. Aileron control tube (Item 278).
2. Slot operating arm (Item 279).
3. Elevator-control crank (Item 280).

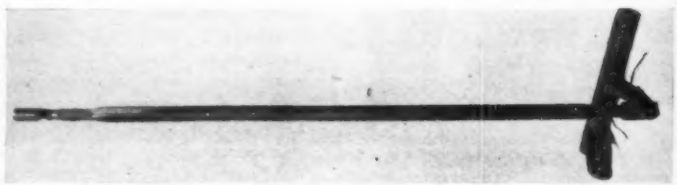


Fig. 33.—Me. 210 aileron control tube.

Fig. 34.—Me. 210 slot operating arm.

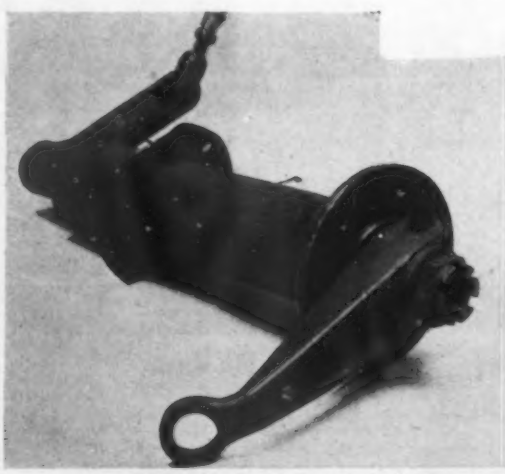


Fig. 35.—Me. 210 elevator control crank.

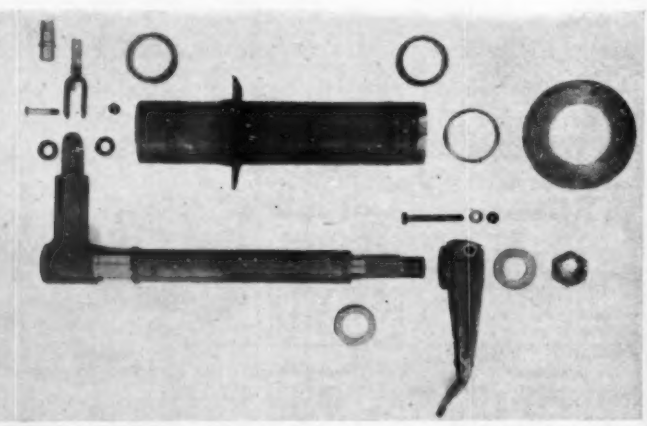


Fig. 36.—Details of Me. 210 elevator control crank.

1. *Aileron Control Tube.*

This component is shown attached to a crank in Fig. 33. The tube proper and the majority of the subsidiary parts had been made from manganese-chromium-vanadium steel containing approximately 0·25% carbon, 0·9–1·2% manganese, 0·6–0·7% chromium, and 0·1–0·2% vanadium. Most of these parts had been used in the softened and cold-drawn condition, with D.P. hardnesses of 238–288. Each end of the tube carried a hardened and tempered fork attachment. That at the free end had been made from 1% chromium-molybdenum steel, with a D.P. hardness of 315-323; the one at the fixed end was a 0·40% carbon steel with a hardness of 220–231.

The crank had been built up by welding, and exhibited normal as-welded structures. Each end of the crank carried a self-aligning two-row ball race, made from a normal carbon-chromium type of steel.

2. *Slot operating Arm.*

The assembly shown in Fig. 34, consisted of a tubular arm loosely riveted to a curved plate, in which were two machined slots. In each slot was a spindle carrying a self-aligning ball race at each end.

The slotted plate had been made from 1% chromium, ¼% molybdenum steel. The range of hardness from surface to core was 328–286. The outer hard portions had a normal hardened and tempered micro-structure, whereas the core showed considerable separation of ferrite. The plate had probably been die hardened to prevent warping.

Races and balls from the various ball bearings were of a fairly normal type of carbon-chromium steel, and the tube and most of the subsidiary parts were made from a manganese-chromium-vanadium steel containing approximately 1·0% manganese, 0·7–1·0% chromium, 0·1–0·2% vanadium, and 0·26–0·29% carbon. Apart from the cold-drawn tube and welded sheet fork attached to the tube and plate, all these parts had been hardened and tempered. The welds in the tubular member showed normal as-welded structures.

3. *Elevator Control Crank.*

The assembly is shown in Fig. 35, and the dismantled parts in Fig. 36. It consisted essentially of two concentrically-mounted tubes. The inner tube, supported by ball races carried in extension pieces welded to each end of the outer tube, also had extension pieces welded to each end. One end carried a welded arm, the other end being splined and supporting a detachable light-alloy arm.

The majority of the parts were made from 1% chromium, 0·2% molybdenum steel containing 0·25–0·30% carbon. The ball bearings were of a normal carbon-chromium type. The various welded joints showed as-welded structures.

Section XIV.—Miscellaneous Parts

THIS section deals with the metallurgical features of a number of components which do not properly belong to any of the groups of parts discussed elsewhere in this series.

A. Mauser 20 m.m. Cannon (M.G. 151/20), from Focke Wulf 190. Aircraft (Item 172)

The cannon was in a damaged condition when removed from the crashed aircraft. A general view is shown in Fig. 1. Examination of the various components of the cannon showed that steels with a wide range of composition had been used, as shown by the table.

The plain carbon and low-alloy steels had been used mainly for the structural and less important parts, and for the light coil springs. The many parts of case-hardened 3½% nickel-chromium steel came from the bolt and breech mechanism, and had been satisfactorily treated, but in several instances the case depth and, consequently, the surface hardness was variable due to unequal grinding.

The high chromium steel was used for four small parts of the cushion brake, and 4½% Ni–1¼% Cr steel was used for the extractor (stamping), and for three components machined from bar, and forming the rear buffer spring.

The barrel was machined from rolled bar of good quality basic electric steel containing approximately 0·3% carbon, 1·5% chromium, and 0·25% molybdenum. It had been heat-treated to possess a tensile strength of 53 tons/sq. in., an elongation value of 28·5%, and Izod impact values of 95–100 ft./lb. The breech end had been carburised to a deptn of 0·03 in., and in addition, it had been locally hardened over 1½ in. of its length, resulting in a surface hardness of 700 D.P.H. and a core hardness of 550 in this region.

The main portion of the breech-block head had been made as a stamping and case-hardened. The grain flow was transverse (Fig. 2), and was presumably designed to strengthen the lugs.

MAUSER CANNON.

Type of Steel	% C	Ni	Cr	Mo	V	Number of Parts
C. H. 3½% Ni–Cr	0·2	3·5	0·75	—	—	12
1% Cr	0·4–0·5	—	1·0–1·5	—	—	2
1% Cr–Mo	0·3–0·45	—	1·0–1·5	0·15–0·25	—	2
1% Cr–V	0·45–0·5	—	1·0	—	0·15–0·25	4
4½% Ni–1% Cr.	0·3–0·4	4·5	1·25	—	—	4
13% Cr	0·2–0·45	—	12–14	—	—	4
Plain Carbon	0·05–0·75	—	—	—	—	35
Spring Steels	**% C**	**Si**	**Cr**	**Mn**		
Si–Mn–Cr	0·6	1·25	0·6	1·0	—	1
3% Si–Mn	0·6	3·0	—	0·75	—	1

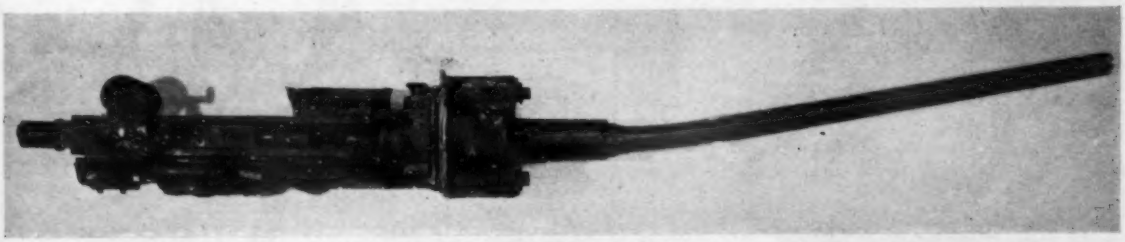


Fig. 1.—Mauser 20 m.m. cannon.

Fig. 2.—Transverse grain flow in stamping from breech block head of Mauser cannon.

B. Armour Plate

The samples examined consisted of a pilot's seat from a Focke Wulf 190 aircraft (Item 179), as shown in Fig. 3, and three pieces from a Messerschmitt 210 aircraft (Items 276 and 277). Three types of steel were used for the four parts, heat-treated to give tensile strengths of 89–109½ tons/sq. in. Details of the dimensions, compositions, etc., of the samples are given in the following table.

ARMOUR PLATE

	F. W. 190 Pilot's Seat	Me 210 Sample 1	Me 210 Sample 2	Me 210 Sample 3
Dimensions, in.	19 × 15 × 0·32 (8 mm.)	22 × 9 × 0·39 (10 mm.)	— (5½ mm.)	13 × 3 × 0·22 (5½ mm.)
Carbon	0·39	0·24	0·41	0·21
Silicon	0·39	0·32	1·65	0·37
Manganese	0·50	1·01	0·93	0·99
Sulphur	0·012	0·026	0·007	0·15
Phosphorous	0·012	0·013	0·011	0·17
Nickel	0·11	Trace	0·03	0·01
Chromium	1·52	0·17	1·17	0·40
Molybdenum	Trace	0·21	Trace	0·21
Vanadium	Nil	Nil	Nil	Nil
Max. Stress T/in.²	95·9; 103·7	93·5; 98·5	109·5; 108·0	89·0
Brinell, H.	467–486	415–444	460–477	364–415

Fig. 3.—Armour plate, pilot's seat, FW. 190 aircraft.

Four mild-steel lugs were arc-welded to the pilot's seat armour. The weld metal appeared to be a high nickel-chromium alloy with 1–1½% molybdenum. No heat-treatment had been given to the component after welding.

C. Seat Support, Dornier 217E–4 Aircraft. (Item 286)

The two portions of the support examined are shown in Fig. 4. They were built up by welding 19-gauge steel tube into "A" form. The main tubes were continuous, the cross tubes being welded to them. A fork attachment was fitted into a threaded sleeve welded to the free end of each main tube, and a clip was welded to the end of the cross tube. A stiffening plate had been introduced into the acute-angled joint formed by the two main tubes.

With the exception of the stiffening plate and cross tube clip, all the essential parts of the support had been made in chromium-molybdenum steel containing approximately 1% chromium, 0·2% molybdenum, and 0·20–0·27% carbon. The stiffening plate contained 0·2% carbon and 1·5% manganese, and the clip contained 0·15% carbon, 1% manganese, 0·75% chromium, and 0·10% vanadium. All the steels were of low sulphur content, and probably made by the basic electric process.

The tubes had been cold drawn to 50–55 tons/sq. in. maximum stress, and the attachment forks and sleeves had been hardened and tempered. All the joints showed normal as-welded structures.

D. Bomb Release Unit, Dornier 217E–1 Aircraft. (Item 131)

The unit is shown complete in Fig. 5, and opened to reveal the working parts in Fig. 6. It was normally electro-magnetically operated, and also carried an auxiliary mechanical release, operated by the plunger seen on the right in the illustrations.

The bomb hook was cadmium plated, and had been machined from a drop-stamping in a plain 0·27% carbon steel. It had been fairly quickly cooled from a high finishing temperature, and

Fig. 4.—Seat support, Do. 217 aircraft.

Fig. 5.—Bomb release unit, Do. 217 aircraft.

flame-hardened at the trigger end to a diamond pyramid hardness of 481. Tensile and impact tests were carried out with a Humfrey Tensograph and miniature Impact Tester on sections from the hook, cut in a longitudinal direction with respect to the grain flow. The following results were obtained.

Maximum Stress, tons/sq. inch	46·0
Elongation on $4\sqrt{A}$, % ..	35·4
Reduction of Area, %	62·0
Miniature Impact Value, ft. lb.	2·67
Approx. Equivalent Izod Impact Value, ft. lb..	51

The bomb hook bush was made from a plain 0·13% carbon steel. It was in the normalised condition, had a fairly uniform D.P. hardness of 172–182, and was lightly cadmium plated over a fine machined finish. Four small bearing bushes were made from a medium carbon steel, normalised and lightly carburised on the working faces to a depth of approximately 0·010 in., the surface hardness being 450–550 (D.P.H.).

High carbon steel (0·68–0·92% carbon) was used for all springs, which were in the hard-drawn patented condition, and gave D.P. hardness values of 525–576.

The housing of the unit consisted of two die-castings made from an aluminium-silicon alloy containing 2·3–4% silicon.

E. Bomb Door Hinge, Dornier 217E–4 Aircraft. (Item 287)

The hinge is shown dismantled in Fig. 7. The mounting shown on the right was flat, and carried two welded ribs, each fitted with a screwed sleeve at its free end. The other mounting plate was shaped and had a single rib welded to it, fitted at its free end with a sleeve carrying a self-aligning ball bearing. The ball races and formed steel cage of the bearing are shown in Fig. 8. The bearing was held in position by two ring springs at each end, these fitting into grooves in the bearing housing. A pliable washer held between each pair of ring springs probably served as a grease-retaining pad.

Most of the parts were made of 1% chromium 0·2% molybdenum steels containing 0·25–0·35% carbon. The shaped mounting plate and the attached sleeve were plain carbon steels. The ball races and balls were carbon-chromium steel. The D.P. hardness of the balls was 946–975 at the surface, and 874–909 at the centre.

The flat mounting plate, including the lower portion of the ribs, had been made as a forging, otherwise the hinge was built up substantially from sheet by welding. The weld attaching the bearing housing to the single rib of the shaped plate appeared to have been heat-treated, but all other joints were in the as-welded condition.

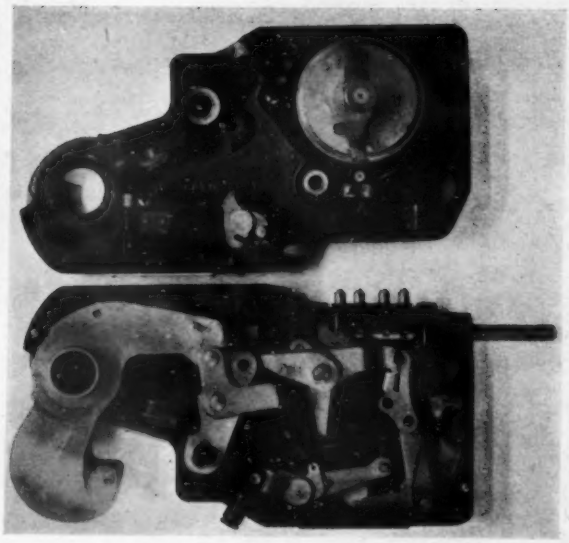


Fig. 6.—Working parts of bomb release unit, Do. 217.

F. Cockpit Window Framework

Samples of the window framework from two Junkers 88 aircraft (Items 106 and 114), and one Messerschmitt 210 aircraft (Item 281) were examined. The framework was in each case built up substantially from welded tube. Both Ju 88 samples were made of 1% chromium, 0·25% molybdenum steel, one being in the annealed and cold-drawn condition with a D.P. Hardness of 228, the other being hardened and tempered to a D.P.H. of 207.

One of the tubes from the Me 210 framework was similar to the above, the others being made from a steel containing approximately 1% manganese, 0·7% chromium, 0·2% vanadium, 0·07–0·15% molybdenum, and 0·23–0·26% carbon. The various tubes were

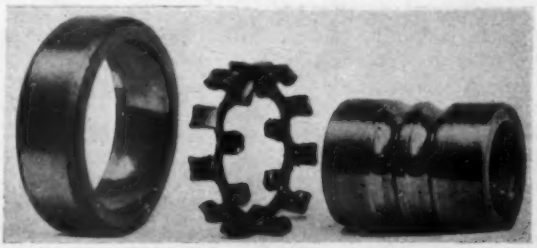


Fig. 8.—Dismantled bearing, Do. 217 bomb door hinge.

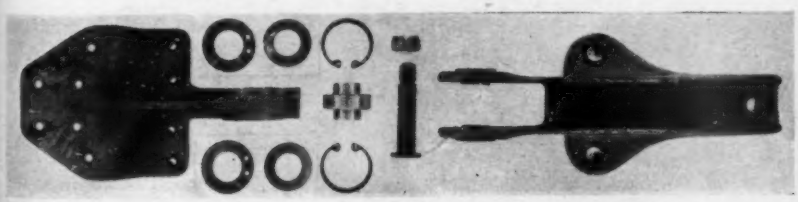


Fig. 7.—Dismantled bomb door hinge, Do. 217.

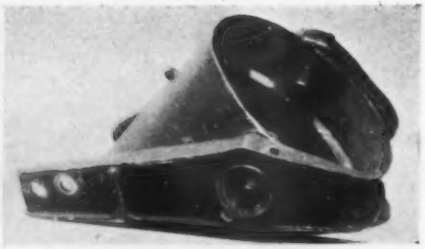


Fig. 9.—Cockpit gun mounting, Ju. 88 aircraft.

either cold drawn or normalised, with D.P. Hardnesses of 191-245.

G. Cockpit Gun Mounting, Junkers 88 Aircraft. (Item 114)

The gun mounting is shown in Fig. 9. It consisted of a cylindrical body with two swivels welded in. A fulcrum pin in each swivel was held in place by a locking screw and a retaining spring. A side arm was placed loose on each swivel. The body, side arm and swivel had been made from 1% chromium, 0·25% molybdenum steel. The body and swivel had been hardened and tempered, and possessed a D.P. Hardness of 219–224. The side lever had been hot worked and softened to a hardness of 166. The fulcrum pin had been made from a plain 0·45% carbon steel, and normalised. It had a D.P. Hardness of 171. All the steels used in the various parts of the mounting appeared to have been made by the basic electric process.

H. Attachment Brackets

Two attachment brackets from a Focke Wulf 190 A-2 aircraft have been examined, as follows:

1. *Wing Rear Attachment Brackets.* (*Item* 175)

The appearance of this component after removal of paint is shown in Fig. 10. It consisted of two parts, a forked bracket and a plate, joined together by arc welding. The forked portion contained two locating holes in each of which was a non-ferrous bush. The forked portion had been made as a forging in a 1% chromium-molybdenum steel containing 0·40% carbon, heat-treated to 62 tons/sq. in. tensile strength.

The plate portion had been rolled from 2% manganese steel containing 0·17% carbon, and heat-treated to 45½ tons/sq. in. tensile strength.

The cleanness and microstructure of both parts were good.

2. *Attachment Bracket for Wing and Engine.* (*Item* 176)

The sample examined was badly damaged, and is shown in Fig. 11. It consisted of a part carrying two lugs, to which were welded two channel section limbs. The lugs were bolted to the upper boom of the centre-section spar, and to one of the engine mounting attachments. The vertical channel section formed part of the framework surrounding the engine bulkhead, and the horizontal limb formed a longeron extending backwards to the rear of the cockpit.

The two channel section members had been made from 2% manganese steel containing 0·14–0·19% carbon. They were in the hot-worked and softened condition, and possessed a D.P. Hardness of 215–228. The portion containing the lugs had been made from a 1% manganese 0·6% chromium steel containing 0·29% carbon. It had been hardened and tempered to a D.P. Hardness of 312–319. One of the lugs carried a ring for a self-aligning bearing. A conventional 1% carbon, 1% chromium steel, hardened and tempered to 736–775 D.P.H., had been used for this ring.

All the steels used had been made by the basic electric process. Welding had been satisfactorily carried out.

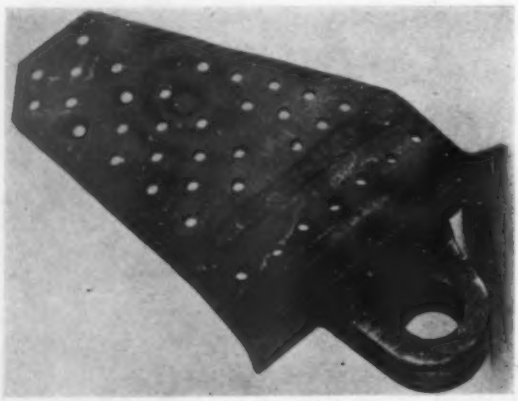


Fig. 10.—Wing rear attachment bracket, FW. 190.

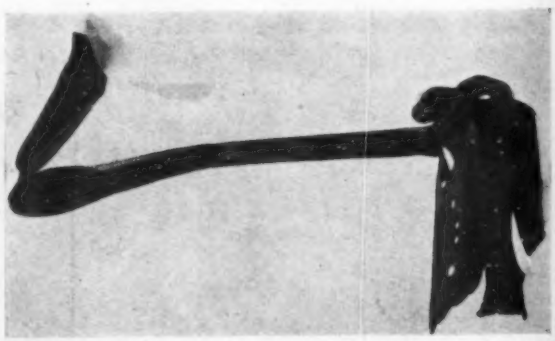


Fig. 11.—Engine and wing attachment bracket, FW 190.

Separation of Cadmium from Zinc by Electrochemical Means

By W. H. Hammond

VARIOUS researches to find a specific analytic method for determining cadmium in the presence of zinc are surveyed. Wet chemical methods are believed to be cumbrous and to lack specificity. Attention is particularly directed to more recent work with the spectrograph and polarograph which indicates that more rapid, sufficiently precise and specific methods are now available for determining cadmium in the low range—i.e., for the spectrograph, 0·0005–0·01%, and for the polarograph (by taking a larger sample upon the low side, or by diluting a smaller sample upon the upper), any amount of cadmium to be expected in slab or die-cast zinc.

For larger amounts of cadmium, such as are found in some silver solders, controlled cathode potential separation is recommended and experimental determinations of cathode potentials of cadmium, zinc and hydrogen, in various baths and under various conditions, are presented.

From Symposium on "Before and After Plating," *Electrochem. Soc.*, 1945, Preprint, 88–27.

Alloy and Tool Steel Manufacture in India

Developments During World War II

By P. H. Kutar, B.A., B.Sc., M.S., Met.E. (U.S.A.)

Modern warfare, because of its enormous and rapid consumption of materials, acts as a stimulus to the development of production methods and tests the ingenuity of scientists and engineers to the utmost. The first world war showed the importance of Indian steel industry as a factor in strategy, this has been greatly emphasised in the recent war during which the Indian steel industry has made great technical advances. This was indicated by Mr. Kutar in his Presidential Address to the Engineering and Metallurgy Section of the recent 33rd Indian Science Congress, held at Bangalore, in which he gave an account of the achievements of the steel industry in India from which the following is abstracted.

DURING the years immediately preceding the World War II., several types of low-alloy high-tensile structural steels were developed in England, America and Germany. These steels were developed to meet the specific requirements of a structural steel, which possessed a higher yield point and higher tensile strength than the ordinary plain carbon structural steel. Such steels had been obtained by the addition of small quantities of alloying elements like chromium, nickel, copper, manganese, phosphorus, silicon, molybdenum, etc. These elements harden the steel, which being low in alloy content, can be conveniently made in the basic open hearth furnace.

The impetus to embark on the manufacture of a suitable alloy structural steel in India presented itself when the project for the construction of the new Howrah Bridge over the River Hooghly was being actively considered. The final specification called for the use of high-tensile structural steel having a tensile strength between 37 and 43 tons/sq. in. The Tata Iron and Steel Co. Ltd., launched a comprehensive programme of research to produce a steel that would comply with the requirements of the specifications in the building of this Bridge. Many difficulties were encountered in the initial stages but ultimately a low alloy steel was developed with the alloy additions of chromium, manganese and copper, which was given the name "Triscom." This was the first high-tensile structural steel produced in India and the new Howrah Bridge in Calcutta was fabricated almost entirely with this steel.

The earlier high tensile steels were developed in order to employ higher fibre stresses in structural design. The range of working stresses involved the use of steel compositions which set a limitation to the possibility of fabrication by welding. During the welding operation, the heat generated tends to produce a hard and brittle zone adjacent to the weld that would endanger the safety of the whole structure. Though safe welding of the "Triscom" type of steel is possible if welding is preceded by heating or succeeded by heat-treatment, it is not practicable to submit large structures to such treatment. It was necessary therefore, to develop a structural steel, which, while having a high yield strength, would lend itself to easy fabrication by welding. After careful investigation a steel, now known as "Tiscor" steel, was developed which satisfied these requirements. This steel contains additions of chromium, copper, silicon and phosphorus and in its development the factor of weldability was uppermost.

Operational difficulties associated with "Tiscor" steel were successfully overcome and, since it has a high yield strength, it has been found to be an ideal structural material for use in the building of freight cars, trolley buses and various other tracting units where reduction in dead weight and an increase in pay load are desired. The applications of "Tiscor" in ship construction may include most structural members of hulls, such as outside plating, interior framing bulk-heads, decks, tank tops; also water-tight doors, hoppers, coal bunkers, ash-wells, sky-lights, boiler breechings, air pre-heaters, stacks, and rudders, etc. This steel has also been found to be more corrosion resistant than ordinary plain carbon structural steel, which is a factor worthy of consideration in the construction of certain structural members.

Bullet-Proof Armour Plates

"Triscom" and "Tiscor" steels were developed in the early thirties. From this time onwards up to the outbreak of the World War II., a small number of electric steel melting furnaces, which are necessary for the manufacture of high quality alloy and tool steels were installed by the Tata Iron and Steel Co., a few other steel makers and by the ordnance factories in India. These furnaces, however, were not put to their rightful use, for they were mainly employed for re-melting scrap and manufacturing plain carbon steels for sections and castings. A few of these steel makers, and the Tata Iron and Steel Co. in particular, did utilise these furnaces to make small quantities of certain alloy steels for their own work's requirement, and it was this experience that stood them in good stead when called upon to supply various qualities and grades of special alloy and tool steels during the recent war period.

In early 1940, the Tata Iron and Steel Co. at Jamshedpur, were asked to supply armour plates for building armoured carriers for service in the Middle East. Having had no prior experience in the manufacture and production of such a class of ordnance steel, the work undertaken had to start practically from first principles. The predominant factor which influenced the selection of the composition and class of steel to be ultimately employed in the manufacture of these plates concerned the facilities afforded by the existing plant and equipment available. Oil-hardening and water-hardening steels were ruled out because these would involve new installations of big and heavy quenching plant and straightening machines. Efforts were, therefore, directed towards the production of an alloy steel of the air-hardening type.

After several experiments on different compositions such a steel was produced on a laboratory scale from

the 75 lb. high-frequency furnace at the new Research and Control Laboratories at Jamshedpur. Samples were rolled and submitted to tests governed by standard ordnance specifications. The initial efforts proved fully successful and were regarded by the Master General of Ordnance Branch, Simla, as excellent and up to home specification. The steel developed was of the air-hardening type and the alloy additions comprised nickel, chromium and molybdenum. The nickel content being 4·0–4·5%. The severe ballistic test requirements were met by a simple normalising treatment followed by tempering at a low temperature.

This steel was then put into commercial production and its manufacture was initially carried out from steel produced in the electric arc furnace ; with the experience gained subsequently it was also produced in the basic open-hearth furnace. The plates were successfully rolled in the existing plate mill, the only one of its kind in India. Close metallurgical control was necessary at all the intermediate stages, from steel-making onwards to rolling, annealing, machining, gas-profile cutting, hardening and tempering, and so on to the finished plate before it was passed on for assembly on the carriers. All the furnaces for the necessary heat-treatment of these plates were built locally and soon a plant was in operation with a capacity of about 800–1,000 tons of annealed plates for fabrication per month.

Of the many difficulties overcome in the successful production of these armour plates, the one which gave the most trouble was the extreme sensitivity of this steel to heating and cooling and the tendency for the occurrence of " flakes." This trouble was eventually cured by a rigorously controlled heating and cooling cycle at the different stages of production.

The early armoured carriers, known as " Tatanagars " were designed for fabrication by riveting and a separate research was necessary to develop a special bullet-proof steel for the manufacture of these rivets which had to be driven hot. It was considered that an air-hardening alloy steel, which could be hardened sufficiently by normalising, at the same time possessing sufficient ductility without the necessity of a tempering treatment, would successfully meet the requirements. This did not present grave problems, as it was found possible to duplicate the armour-plate composition with the difference that the carbon and chromium contents in the steel (the two principal carbide forming elements and therefore conducive to hardness) were maintained at much lower levels. A successful bullet-proof steel for rivets was thus produced, which was subsequently used for the manufacture of bullet-proof rivets for the armoured carriers.

As the " Tatanagars " started rolling out of the assembly line in large numbers, it was soon realised that the shearing and riveting of the plates were causing a regular bottle-neck in production, and, therefore, gas profile cutting of plates and a riveted-cum-welded construction was suggested through the initiative of Tatas. This suggestion was put forward after tests were carried out on riveted and welded joints, whereby it was proved that the latter were much stronger than the former. A riveted joint, moreover, suffered from a grave disadvantage in combat service in that if an armour-piercing bullet happened to strike on the head of a rivet, the occupants inside the carrier were liable to be seriously hurt by the splashing of the metal from the rivet shank. In a welded construction this danger did not exist as the metal joint was continuous and the weld was also bullet-proof.

Arc Welding of Armour Plates

There were certain major difficulties in the way of adopting the method of all-welded construction. The lack of suitable welding equipment and competent welders at the time, were factors which stood in the way of the adoption of the all-welded construction. In order to expedite the manufacture of the armoured-carrier bodies, it was decided, therefore, to adopt a riveted-cum-welded method of fabrication. Gas cutting of the armour plates did not present too difficult a problem, as it was soon found that the tendency towards cracking at the edges of parts, cut from plates of heavy sections, could be largely avoided by an initial annealing treatment. On the other hand, the welding of bullet-proof steel with marked air-hardening characteristics raised problems affecting the composition of the plate, the welding rod and the welding technique.

The problem of composition of plate was solved to a large extent by reducing the hardenability of the steel by lowering its carbon content and thus decreasing the brittleness of the weld-hardened zone, but sufficient carbon was retained so that the ballistic properties would not be impaired. Since imported supplies of welding electrodes were scarce and irregular, experimental work was carried out with a view to producing suitable flux-coated electrodes in India and ultimately a special type of electrode was developed. The development of a suitable welding technique also called for careful study and after a series of methods had been tried and the resulting welds submitted to searching tests, to ensure that the joints had sufficient strength and toughness, from which a suitable method for welding the air-hardening steel emanated.

Evolution of a New Composition Armour Plate

The chequered development of the manufacture of bullet-proof armour plates in India had to undergo another vicissitude in its short history when the battle of the desert was being fought to a standstill and shipments of nickel to this country were seriously threatened. Nickel was in short supply, and the necessity for the conservation of the existing stocks of this element was made imperative. Further research became necessary, as a result of which a new composition for armour plate was developed possessing the good ballistic properties and welding characteristics comparable with the earlier composition. In this new composition the nickel content was cut to one-third of that of the original composition. Experiments were carried out to determine the most balanced combination of the various alloying elements, until most of the nickel in the original composition had been replaced by copper and silicon ; the chromium and manganese contents were increased to maintain the ballistic properties at a high level.

Synthetic Hæmatite Pig Iron

Previous to the war the ordnance factories of India followed the general practice of mixing special scrap and hæmatite pig-iron for acid steel manufacture. Soon after the outbreak of hostilities, however, supplies of hæmatite pig-iron were cut off and the ordnance factories were placed in a serious dilemma. After careful investigation it was found that an all scrap charge, with the addition of carbon in the form of petroleum coke, satisfied requirements and a new procedure was developed. The synthesis was carried

out in the orthodox manner by melting steel scrap and an excess of coke in a basic electric arc furnace of the Heroult type and the iron so made was cast into pigs. Being low in phosphorus and sulphur contents, this iron proved to be an ideal substitute for hæmatite pig for use in making acid steel as well as a recarboniser for addition in the manufacture of high-quality alloy and tool steels. Later a process was developed for making this synthetic pig iron in the basic open-hearth furnace which was cheaper than but as good as that made in the electric arc furnace.

Permanent Magnet Steels

Research on permanent magnet steels was undertaken to meet the urgent demands of the Posts and Telegraphs Department of the Government of India, due to acute shortages of imported magnet steel bars. After much study and research, several types of magnet steels were successfully produced, including the chromium, tungsten and the high cobalt types. These steels are not easy to make and even more difficult to roll or forge. Eventually, however, the correct methods of processing were evolved to obtain the best results. The service performances of these steels have been reported to compare favourably with the imported steels.

High-Silicon Steel Sheets for Electrical Machines

During the war, several enquiries had been made from time to time for the supply of high-silicon steel sheets for use in the manufacture of electrical motors, dynamos and transformers. Since imports had dwindled down during the war, Indian industry was approached to fill in in the gap. The manufacture of this steel did not present great difficulties. The rolling, however, and the subsequent processing of the sheets did require a considerable amount of work before finally a correct technique could be evolved to produce a sheet that would possess the necessary magnetic and electrical properties. The rolling and finishing temperatures on high-silicon sheets are critical. Similarly, the annealing cycle is even more important and should be closely controlled, so that a critical grain size and a definite orientation is obtained to give the best properties. These factors are not easy to control and the difficulties were enhanced because of the necessity of working with the existing plant and equipment which, in most cases were not quite adequate to handle the new work. Despite the difficulties, however, improvisations had to be effected and the work was eventually carried through successfully. The production of these high-silicon steel sheets in India has helped to bring about the prospect of establishing a stamping works in this country where these sheets would be used for stamping out laminations and other forms for use in the assembly of electric motors, dynamos and transformers.

Miscellaneous Tool Steels

During the war, when the shortage of small coins was being acutely felt, a general expansion of the production capacities of the existing mints was planned by the Government. It was soon realised that a sufficient stock of imported die steel was not available to meet the increased demand. A careful study was made of the type of steel needed, and soon, a steel was offered to meet the necessary requirement of the mints in this respect. The supply of mint die steel may be said to be the first attempt at the manufacture of quality tool steel in India. The early attempts did not meet with entire success as the dies did not reach the same standard of performance in service as those of foreign manufacture. As more experience was gathered, however, the processing of the steel was improved upon, and very satisfactory service performances have since been reported.

Enquiries continued from various sections of industry for tungsten die steels for hot work, high-carbon high-chromium steels for extrusion dies, low-tungsten steels for hack-saw blades and oil-well bullets, medium-tungsten high-carbon steels for pneumatic tools, low-chromium high-carbon steels for razor blades, etc. These demands have been met with steel manufactured in the electric arc furnace and the high-frequency induction melting furnace.

Corrosion and Heat-Resisting Steels

The manufacture of these steels in India had never been attempted, but, since imported material was in short supply, ways and means had to be found to meet urgent demands. Eventually these steels were successfully produced in the electric arc furnace and the high-frequency induction furnace. Stainless steel bars have been supplied for the manufacture of surgical instruments ; and sheets for the manufacture of acid-resisting vessels for chemical plants. Heat-resisting steels for resistance to scaling at high temperatures, which contain high percentages of chromium and nickel have also been manufactured and supplied for steam valves and valves for internal combustion engines.

High-Speed Tool Steels

Of all alloy steels manufactured, no other group of alloy steels has greater strategic value than the tungsten high-speed steels. In the early days of the war, when the prospects of importing these steels was bleak, the Tata Iron and Steel Co. were approached by the Government to undertake the manufacture of high-speed steels. A bigger furnace, 10-cwt. capacity, was made available for installation at the works of Tisco at Jamshedpur for this purpose. Fortunately, the Company had sufficient experience of this type of steel before the war, but India had been dependent on imports of ferro-tungsten and this presented a serious problem. However, this problem was tackled with determination. Wolfram ore obtained from Judhpor and elsewhere was reduced in electric arc furnaces designed and built locally and soon a pilot plant was in operation for the vital alloy, ferro-tungsten. The production of high-speed steel was beset with difficulties, not because of lack of knowledge or experience, but, primarily, because of insufficient forging capacity. All efforts were made to increase the capacity in this direction and eventually it was possible to meet all the urgent demands of the Government for this grade of tool steel. The development of high-speed steel produced in India had to pass through great vicissitudes and adverse criticisms at the hands of indentors and users. These steels were subjected to repeated practical tests to compare their performance with similar standard grades of imported high-speed steels, and the locally made steels in no case suffered by comparison.

The production of high-speed steel still continues in this country, and one may well be proud of our achievement, for had it not been possible to supply this steel at a time when it was most needed in this country, much of the industrial war machine in India would have been at a stand-still and the vital war supplies from the " Arsenal of the East " would have been very seriously

jeopardised. In almost every case, production manufacture had to be preceded by carefully planned laboratory-scale experiments in order to indicate to the consumer the correct processing to be adopted during subsequent manufacture of the finished article. To describe the difficulties encountered in the manufacture of the above steels would be beyond the scope of this short review. Suffice it to say that the difficulties were solved successfully and the demands were satisfied despite the fact that all the processing had to be carried out with equipment that had to be improvised, as the manufacture was carried out in plants which were not fully equipped to undertake the manufacture and production of these special types of steels.

It is apparent that in the field of alloy-steel manufacture in India, great strides have been made, and the experience gained during the last few years should stand in good stead for the future development of this branch of the steel industry. The age-old myth that it is necessary to acquire half a century's experience before quality tool steels can be made, may be said to have been exploded for all time, when it is realised that it was possible to develop and produce armour plate in India on a commercial scale within the short period of about four months. Science has moved forward in the last twenty years. The pace of development has also accelerated, and facilities that did not exist twenty or fifty years ago, are now available for development. Our ways of thinking and working, therefore, have to be recast and made to follow rational as well as practical lines in keeping with modern trends and developments.

Relationship of the Geological Survey Department to the Mining Industry of Malaya

In the third of the series of lectures on the mineral resources of the Colonies delivered on February 28 last, at the Imperial Institute, Mr. E. S. Milburn, Director of the Geological Survey for the Federated Malay States, first recounted the achievements of the Malayan tin industry, the most important of its kind in the Colonial Empire, and then turned to geological factors controlling the location of the tin deposits and thence to the work of the Geological Survey Department since its inception in 1903.

Exports of metallic tin from 1900 until the Japanese invasion, not counting those from the Unfederated States, amounted to more than two million tons, or nearly 40% of the world's output. In 1939, while restriction was in force, the Federated Malay States produced 52,232 tons of metallic tin or 28·3% of the world's output. With the abatement of restrictions, 41,585 tons were produced in the first half only of 1941. Not very much tin was mined in Malaya during the Japanese occupation.

Since it was first mined in 1915, 10,000,000 tons of coal had been won. In the last 50 years 800,000 oz. of gold had been mined. Prior to the occupation, high-grade iron ore had been exported to Japan at the rate of over one and a half million tons a year, bauxite at 55,000 tons p.a., and manganese ore at 30,000 tons p.a. Until the famous Kramat Pulai scheelite mine closed in 1939, the annual rate of tungsten exports was a matter of thousands of tons, and it continued at the rate of hundreds of tons until the Japanese attacked.

The value of the tin produced since 1900 was of the order of £500,000,000 and the value of the other minerals produced during the same period might have been £50,000,000.

Discussing the geological factors controlling the location of the tin deposits and also the work of the Geological Survey Department, Mr. Milburn gave a vivid picture of the exceptional difficulties attending geological surveying in Malaya. In the first thirty years or so a reconnaissance survey of the country was made. The main object of the Department, as now constituted, was to prepare and publish detailed maps of those areas indicated by the reconnaissance survey as most likely to afford mineral deposits. In the past, little deep prospecting had been carried out by the Geological Survey, though at the time of the Japanese invasion two areas were being drilled under the Department's direction, one for tin and the other for coal. In future, any areas found by a geologist during the course of detailed mapping that he considered needed testing would be prospected.

Mr. Milburn also stressed, with examples, the valuable geological information that had been yielded by the Survey's continual study of working mines, which also helped to ensure that known mineral deposits were mined to the best advantage. Another of the Survey's activities was the collection of all prospecting results, old as well as new. Thus for some years, any person proposing to prospect in any part of Malaya had been at liberty to consult this collection to see what information was available about the area that interested him, due regard being paid to existing mining rights.

By using the mass of information now collected, the Geological Survey had prepared maps dividing the country into the following categories: (*a*) Mining land; (*b*) potential mining land; (*c*) land where mineral deposits might be found; and (*d*) land that could be reserved for purposes other than mining. The ideal would be, of course, to divide the whole country into the first and last only of these categories and it was hoped, in time, as the necessary information became available, to achieve this object.

Statistical Bulletin of the British Iron and Steel Federation—Vol. 21, No. 1

With this issue the British Iron and Steel Federation resumes the publication of the Statistical Bulletin suspended before the war. Since the end of the war publication of statistical information has been resumed, first with the issue of comprehensive statistics bringing the annual United Kingdom figures of 1938 issue fully up to date to the end of 1944. In addition, monthly figures of pig-iron and steel ingot production have been published for 1945. This issue of the *Monthly Bulletin*, however, carries the monthly United Kingdom figures into much greater detail, and includes particulars for certain other countries. It is proposed to develop further the information regarding other countries as quickly as reliable particulars become available.

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The Institute of Metals

Annual General Meeting

The Thirty-eighth Annual General Meeting of the above Institute was held at the Institution of Civil Engineers, London, on March 13 and 14, 1946, at which the number of members present approached nearly to pre-war standards despite difficulties which continue to be experienced in obtaining suitable hotel accommodation. Here, it is possible only to summarise the proceedings at the initial meeting and to present a résumé of the papers presented at the technical sessions.

IN opening the proceedings at the above meeting the retiring President, Dr. W. T. Griffiths, commenting on the Report of Council for the year ended December 31, 1945, said we must be gratified to find that membership of the Institute continues to increase. At the end of the year the membership was only two short of 2,600 ; since then many more applications have been received, and members separated from the Institute during the war years have returned, so that the present membership is substantially higher than that given in the Report.

The President referred feelingly to the passing of a number of members, in particular of Engineer Vice-Admiral Sir George Goodwin, a Past-President, and Dr. G. D. Benbough, who took a great part in the deliberations of the Institute.

In his further review of the Report of Council the President referred to the formation of the Institute of Metallurgists and extended most cordial wishes for its successful future. The subject of education, he continued, has taken up a considerable part of the time of Council during the past few years, and, in order to co-ordinate the efforts of all the bodies concerned, a Joint Committee on Metallurgical Education has been formed with the Iron and Steel Institute, the Institution of Mining and Metallurgy, the Institute of British Foundrymen and the Institution of Metallurgists. It is hoped that National Certificates in Metallurgy, which have now been established, will be granted at the end of the current educational session.

Fellowships in Metallurgy

The Hon. Treasurer, Mr. H. S. Tasker, after briefly reviewing the accounts of the Institute, referred to the handsome offer which has been made by the President's company, The Mond Nickel Company, of £50,000 to institute fellowships in metallurgy. The offer has been made to the various metallurgical institutions jointly, and the letter conveying the offer, while in no way attempting to diminish the enthusiasm for increasing research, does emphasise the need for at least equal importance being given to the provision of means by which the results obtained in the research laboratory can be utilised in industry. Both questions, that of increasing research and that of securing the prompt utilisation of research results, are related to a third : that of the availability for executive and administrative posts in industry of men able to appreciate the technological significance of research and its results.

The offer makes available, during these critical post-war years, a number of Fellowships for award annually to individuals of particular promise who, having received an adequate academic training, desire an opportunity of broadening their experience or training. Since it is the years in the immediate future that would seem to matter most, it is suggested that the total sum of £50,000 and any accrued interest should be expended during the next ten to fifteen years. It is estimated that in this way at least five Fellowships of an average value of £700 would be available each year. The Fellowships need not be confined to those who are qualified metallurgically, but could be used to attract into the metallurgical field men qualified in other sciences who might, if unable to get suitable experience, meet a need in the metallurgical industry.

As far as this Institute is concerned, continued Mr. Tasker, a gift of this sort is of the greatest possible value to all of us. It is a very great encouragement to those of us who have been working in the field of metallurgical education ; what is more important, it is a great encouragement to those engaged in teaching metallurgy in various centres ; and, more important still, it is a great encouragement to their victims, the students of metallurgy, some of whom at any rate will benefit directly by this very munificent gift. Mr. Tasker moved and Mr. A. J. G. Smout seconded : "That the members of the Institute in General Meeting express the sincerest gratitude to the Mond Nickel Company and to the International Nickel Co. of Canada for this offer."

This was carried with acclamation.

Election of Officers

Officers elected to fill vacancies on the Council for 1946-47 were as follows : *President :* Col. P. G. J. Gueterbock, C.B., D.S.O., M.C., T.D., M.A. *Vice-Presidents :* Professor G. Wesley Austin, O.B.E., M.Sc., Mr. W. F. Brazener and Mr. A. J. G. Smout.

Members of Council : As a result of a ballot the following four members received the most votes : Dr. N. P. Allen, M.Met., D.Sc., Dr. Hugh O'Neill, M.Met., D.Sc., Dr. J. W. Jenkin, B.Sc., Ph.D., and Professor F. C. Thompson, M.Sc., D.Met.

May Lecture

The President announced that the next General Meeting of the Institute will be held on May 22, probably at 6 p.m., in the Hall of the Institution of Mechanical Engineers. The lecture this year will be given by Professor N. F. Mott, M.A., F.R.S., of Bristol University, who has taken as his subject : "Atomic Physics and the Mechanical Strength of Metals."

Award of Platinum Medal

The Council unanimously decided that the 1946 Medal should be awarded to Sir John Greenly. In making the presentation the President said : Our country, and therefore each one of us individually, owes Sir John Greenly much for his services to the nation during two world wars. In the 1914-18 conflict, after a period of distinguished service overseas he was appointed Deputy Controller of Inspection of the Ministry of

Munitions, and did much to facilitate the increased production of those metal-constituted munitions which were then so urgently needed. Before the recent war, at a time when at last it was realised that preparations for war had urgently to be made, Sir John undertook the heavy responsibility of the Chairmanship of the Prime Minister's advisory panel of industrialists. When war broke out, he went to America to take charge of the purchase of war supplies there; and, as we know, for his work during the war, in which certainly his knowledge of the non-ferrous industries must have been of help, and from which those industries benefited, His Majesty in 1941 conferred upon him the K.C.M.G.

Sir John, however, has had a closer contact with the non-ferrous metal industries than might have been expected either of an engineering user of metals or of an industrial organiser. In 1920 he entered the non-ferrous metals industry as Joint Managing Director of the old-established South Wales Copper Smelting Company, and he remained in the industry for several years. One consequence of his activities in this direction was his election in 1925 to the Council of the British Non-Ferrous Metals Research Association, of which he subsequently became Chairman an onerous office which he now holds.

The Institute of Metals and the British Non-Ferrous Metals Research Association are the two bodies which unite the many and diversified elements of the non-ferrous metals industry. In bringing both to their present position, Sir John has played an important role. Amid applause, the President handed the Medal to Lieut.-Col. Sir John Greenly.

Technical Sessions

Following these proceedings technical sessions were held at which some seven papers were presented for discussion, a very brief résumé of which is given in the following notes.

The Intercrystalline Corrosion by Sodium Chloride of a Wrought Aluminium Alloy Containing 4% Copper

By J. D. Grogan, B.A.

IN an earlier investigation[1] it was shown that aluminium containing 4% of copper is liable to intercrystalline corrosion after being cooled in air from the solution-heat-treatment temperature. In that investigation the corroding agent used was sea-water spray. Later, a systematic investigation was carried out on the types of chemical reagents capable of producing intercrystalline attack in this material, and the behaviour of sodium chloride was studied in detail, a report of which is given in the present paper. The work described by the author was carried out under the general research programme of the National Physical Laboratory.

It is shown that when a wrought aluminium-copper alloy containing 4% of copper, cooled in air from the solution-heat-treatment temperature, is exposed to attack by solutions of certain acids—*e.g.*, hydrochloric acid, the incidence of attack falls mainly on the grain boundaries and the material becomes brittle. When caustic soda is the corroding agent general attack occurs, and round shallow pits are formed in the grain boundaries; the material does not become brittle. When the alloy is exposed to sodium chloride solution in the presence of air, oxygen concentration cells may develop. Areas deficient of oxygen become anodic, and here the hydrochloric acid intercrystalline type of attack occurs. Other areas with higher oxygen concentration become cathodic, and here the caustic soda non-embrittling type of attack develops.

The mode of attack of hydrochloric acid on the material quenched in cold water is different and occurs along certain atomic planes of the crystals; whole crystals may be eaten away, while the boundaries remain relatively free from attack.

[1] J. D. Grogan and R. J. Pleasance, *J. Inst. Metals*, 1939, **64**, 57.

Damping Capacity and the Fatigue of Metals

By R. F. Hanstock, Ph.D., and A. Murray.

FATIGUE failures usually occur without warning, and although attempts have been made to detect the commencement and development of fatigue—e.g., by metallographic and X-ray examinations, on the whole they have not been successful. There is, however, one property of metals which may be closely related to fatigue. This is "damping capacity," a measure of the ability of the metal to convert vibrational energy irreversibly to other forms—e.g., thermal energy and potential energy within cold-worked crystal structures. Since fatigue must be attributed to work done on the material by the alternating stresses, there is good reason for suspecting an association between damping capacity and fatigue. That such an association has not yet been demonstrated satisfactorily may result from the conventional methods of investigation.

The research described in the present report was undertaken to develop a method of fatigue testing which would allow determinations to be made of the damping capacity of the specimens during the course of the test. It has involved an extensive investigation of problems associated with the accurate measurement of damping capacity over a wide range of stresses, and with the maintenance of resonance vibrations of high amplitude. As a result of this work, which has been carried out in the laboratories of High Duty Alloys, Ltd., it has been possible to investigate fatigue and damping capacity simultaneously for some aluminium alloys, and there is a reasonable expectation that eventually the method may be applied to a wide variety of materials.

The report is divided into several sections, the first of which describes equipment for the excitation of resonance vibrations and for the measurement of damping capacity; other sections present some of the information obtained by examination of selected materials. Particular attention has been given to certain aluminimum alloys in which fatigue phenomena have been investigated by observing changes in the damping capacity of the material during continuous vibration at surface sheer stresses up to 10 tons per sq. in. A high frequency of stress alternation—of the order of 1,500 cycles per sec.—was employed, the specimen being freely suspended and vibrated torsionally at the resonance frequency. The alternating torque is applied electromagnetically, and the maximum surface sheer stress developed depends on the damping capacity of the specimen.

It is shown that changes in damping capacity of the specimen during vibration at a stress above a critical value are associated with strain-hardening and fatigue, the former being indicated by a gradual decrease, and the latter by a gradual increase, in the damping capacity. Increase in damping occurs as a result of the formation

of fatigue cracks, and at a given vibrational stress it appears that the time rate of this increase may be indicative of the endurance of the material.

As a result of the work done, the authors conclude that the damping capacity of a metal depends very sensitively on the composition and metallurgical state of the material. This sensitivity is fully revealed only when other losses of vibrational energy are eliminated. The extent to which extraneous losses of energy can be reduced, and the necessity for ensuring that such losses are very small, is demonstrated by the values obtained for the damping capacities of a number of alloys. These values are very much lower than had been anticipated from previously-published data, and consequently it is believed that the method adopted is capable of measuring damping capacity with a precision hitherto not obtained over a comparable range of vibrational strains.

The very low damping capacities obtained at quite high vibrational strains raise the question of the practical value of damping capacity as a factor in the control of vibrational stresses at resonance. The values found by the present method of investigation suggest, at least for aluminium alloys, that the intrinsic damping capacity will exert very little control on the development of vibrational stresses below those stresses at which fatigue occurs. In most practical systems the vibrational stress at resonance will be determined by the external damping, which may be far greater than the damping capacity of the material. The relation between damping capacity and stress has, however, considerable practical significance. For aluminium alloys, it has been shown that fatigue does not occur until a critical vibrational strain is reached, which is indicated by a sharp increase in damping capacity. Provided that the external losses in the system are sufficient to prevent an amplitude of vibration corresponding to the critical strain from being developed, long endurance may be expected.

Consequently, for long endurance it may be desirable to select materials having a high critical strain and a low damping capacity at strains less than the critical value. Materials which strain-harden during vibration may, of course, be able to withstand higher vibrational stresses than are indicated by their initial strain, for the magnitude of this will tend to increase during vibration. It is possible that the period during which strain-hardening occurs may be regarded as a "safe" period in the life of the material and in systems having low external damping, the use of such material may be justified, provided the normal operating life is not greater than the "safe" period corresponding to the operational stresses.

The hypothesis that a high rate of increase in damping implies rapid failure by fatigue, will, if substantiated by further investigations, provide a quick method for assessing the fatigue properties of materials. The general application of such a method will, however, involve development of the equipment to obtain efficient transference of considerably more energy per cycle to the specimen. This is not solely a question of increasing the power output of the vibration amplifier; a more important aspect is the development of exciter coils capable of handling an increased supply of power with greater efficiency. It is believed that considerable improvements may yet be made in this direction.

Some Experiments on the Effect of Residual Stresses on the Fatigue of Aluminium Alloys

By G. Forrest, B.Sc.

IT has long been realised that metals and alloys in their finally-manufactured forms frequently contain residual and internal stress systems. These stress systems are a consequence of the method of manufacture employed, and in many cases are unavoidable in the final product. Further, some forms of mechanical processing applied to a material after manufacturing will superimpose additional internal stress systems on any that may exist in the material as manufactured. These internal stresses are caused by non-uniform thermal operations and non-unitorm plastic deformation.

A selection of the most significant results, from investigations on the effects of residual stresses on the fatigue properties of aluminium alloys, obtained in the laooratories of the Northern Aluminium Company's Research Department, is contained in the present paper, the first part of which describes rotating-cantilever fatigue tests on two wrought and one cast aluminium alloy, some of the specimens in each case containing residual stresses due to quenching. The stresses had little effect on the fatigue strength of plain (un-notched) specimens of wrought material, but on specimens having annular notches, increases were obtained of up to 80% above the strength of material not containing residual stresses. Plain (un-notched) specimens of the cast material showed a 20% increase in fatigue strength resulting from the presence of residual stresses.

The second part describes fatigue tests on tubes with external annular notches. The results are discussed in relation to the residual stress system associated with a heavy sinking pass.

In the third part the effect of residual compressive stresses at the root of an annular notch is dealt with, the stresses being induced by pre-loading the specimen in tension. Pre-loading resulted in an improvement of the fatigue properties by 100% in the rotating cantilever test.

The fourth part describes fatigue tests on single spot welds, some of which were pre-loaded in the direction of subsequent application of the fatigue loads. The fatigue strength of pre-loaded spot welds showed an increase of 50% over that of the original welds under fluctuating sheer stresses.

The main conclusion drawn by the author from the results of this work is that, under the complex stress conditions at the root of an annular notch, compression stresses—whether applied by quenching or mechanical action—may result in an extremely large increase in fatigue strength. This is based on the assumption that the increase of strength is in fact the result of the residual stress, and that effects of other differences have been small. Regarding the test-pieces with annular notches initially stressed by loading in tension, two other factors may be considered—*viz.*, the change of shape of the notch, and the effect of cold work. The measurements made on test-pieces before and after pre-loading have shown that the change of shape of the notch was very small. Regarding the effects of cold work on the fatigue properties, a previous investigation carried out in the Northern Aluminium Company's laboratories has shown that, for test-pieces without notches, cold work by stretching after heat-treatment tends to have little effect on the fatigue

strength of extruded Duralumin-type alloys except at high degrees of strain, when there is a small decrease.

The problem is more complicated with respect to spot welds, since the material at the critical part of the joint cannot be regarded as normal Duralumin-type material. Tests reported by Field, Sutton, and Dixon[1] showed that spot welds in material to B.S.S. 5L3 gave an appreciable increase in strength after cold rolling with a 2·7% reduction, whereas a greater reduction resulted in no improvement over the value for unrolled material. The 2·7% reduction was insufficient to cold-work the actual welds, and in this case it appears possible that the increase due to the cold rolling was also a result of redistribution of internal stress in the joint.

Computations of Rolling and, Torque and Roll-Face Pressure in Metal Strip Rolling

By Maurice Cook, D.Sc., Ph.D., and Eustace C. Larke.

IN the cold-rolling of metals the magnitudes of the rolling load, and the roll torque and the distribution of pressure on those parts of the roll surfaces which are in contact with the material being rolled, are dependent upon a number of factors, including the reduction in thickness of the metal being rolled, its initial thickness, the roll diameter, and the frictional forces being generated between the surfaces of the rolls and of the metal. The roll diameter, initial thickness, amount of reduction, and friction conditions are factors which can be altered independently of each other and, collectively, are referred to as the independent factors, in contrast to the rolling load, roll torque, and roll-face pressure, which are, collectively, termed the dependent quantities.

During the last 20 years several workers have developed formulae from which the dependent quantities can be calculated. With one exception, each of the methods for calculating the dependent quantities involves establishing a curve showing the manner in which the pressure varies throughout the arc of contact. This curve is usually referred to as the friction hill, and the rolling load and roll torque are proportional to the area beneath it. Values of rolling load, calculated from roll pressure curves, differed from experimentally determined values, and one of the objects of the present work is to show, in simple graphical form, for the cold rolling of H.C. copper, the effect of variations in the independent factors on the dependent quantities. The ranges over which the independent factors have been varied are: roll diameters 6, 12, 16 and 24 in., initial thickness 0·1, 0·2, and 0·5 in., reductions of 10, 20, and 30%, and coefficients of friction 0·10, 0·15, and 0·20. Concerning the frictional coefficient, the lowest value adopted approximates to conditions resulting from the use of smooth-surface rolls, that is, rolls normally ground and polished, whilst the highest value approximates to conditions resulting from the use of relatively rough rolls, such as those which have been in regular service for an appreciable period of time. Sixty different combinations were used.

The effects of roll diameter, initial thickness, amount of reduction, and the frictional forces encountered between the surfaces of the rolls, and the metal, on the magnitude of the rolling load, roll torque, and roll-face pressure, in the rolling of copper strip, have been determined by applying the Orowan* graphical method of calculation and, where suitable, the formulæ of Ekelund†. The results are presented in the form of diagrams, from which it is possible to read off values of rolling load, torque, and roll-face pressure in the range of conditions examined.

In addition, values of rolling load and torque have been computed in the cold rolling of 70 : 30 brass, mild steel, and an aluminium alloy of the Duralumin type, in strip form, using both methods of calculation for a selected number of the sixty different combinations of rolling conditions considered in the cold rolling of copper.

A comparison of the values of rolling load and torque calculated by the two methods shows that those computed from the Ekelund formulæ, modified by the addition of simple factors for each material, are within about 6% of those derived by the Orowan method.

The Reaction of an Aluminium-Magnesium Alloy with Water Vapour and the Absorption of Hydrogen

By R. Eborall, B.A., and C. E. Ransley, Ph.D.

THE tensile strength of the aluminium-magnesium alloys in the annealed conditions is good, particularly of those containing 5–7% magnesium. The strength may easily be raised to a high value by work-hardening without heat-treatment. These alloys also have good mechanical properties in the cast condition, and they are also suitable for fabrication by welding. Unfortunately, alloys containing over 5% of magnesium are liable to exhibit severe-porosity in the junction zone between the weld metal and the basic sheet, and this effect is frequently more pronounced when material of thick section is welded. Similar considerations have militated against the use of higher-magnesium alloys for sand casting, since considerable porosity occurs unless the sand is inhibited. Some difficulty has also been experienced in the manufacture of sheet and other products, when blistering may occur on annealing after cold work.

The British Non-Ferrous Metals Research Association has been studying these alloys, with particular reference to those features which at present limit their application, an account of which will eventually be published. The Association's unpublished work on the unsoundness which occurs in sand casting and in welding may, as far as it is relevant, be briefly summarised.‡ In the first place, the effects observed, for example in gas welding, were not peculiar to that process, since severe internal discontinuities could be produced merely by annealing in various atmospheres at a temperature between the solidus and liquidus; such treatment caused a reaction to take place which resulted in blistering and heavy oxidation accompanied by a considerable evolution of heat. Secondly, this behaviour was typical of alloys containing magnesium, and was not shown at all by other alloys of similar freezing range—e.g., aluminium-silicon and aluminium-copper alloys. These results suggested that a preferred oxidation was taking place, involving the decomposition of water vapour and the absorption of hydrogen, and this conclusion was supported by the fact that annealing in the critical range was not harmful if carried out in a

1 G. R. Field, H. Sutton, and H. E. Dixon, *Trans. Inst. Weld.*, 1943, **6**, 49.

* E. Orowan, *Proc. Inst. Mech. Eng.*, 1943, **150**, 140.

† S. Ekelund, *Jernkontorets Ann.*, 1927, **111**, 39; 1928, **112**, 67; *Steel*, 1933, **93**, (8), 27; (9), 44; (10), 38; (11), 32.

‡ This work has been described in a B.N.F.M.R.A. report by W. L. Hall and E. A. G. Liddiard; work carried out more recently on welding is dealt with in reports by J. Pendleton.

really dry atmosphere. Attempts were made to inhibit this reaction by producing a protective film on the surface. Of these, the two most successful were the addition of small amounts of beryilium to the alloy, and the introduction of hydrogen fluoride into the atmosphere (as has sometimes been done commercially) during heat-treatments. Unfortunately, however, the fluxes used in welding must remove all such films, so that both these remedies had no effect on the porosity due to the absorption of hydrogen during welding, although additions of beryllium have been shown to be beneficial in sand casting.

It seemed desirable to have some more direct and fundamental information on the absorption of gases by these alloys, and the work described in the present paper was therefore initiated. It was carried out at the Research Laboratories of the General Electric Company, Ltd., using apparatus which had previously been developed there for investigating the gas contents of metals.

Using the apparatus and method described, it has been possible to study :

1. The true gas content of alloys containing up to 7% of magnesium in the form of cast metal or rolled sheet, appropriate correction being made for surface gas.
2. The diffusivity of hydrogen in the solid alloy at various temperatures.
3. The equilibrium solubility of hydrogen in the solid alloy (only preliminary measurements have been made).
4. The reaction with water vapour at 600° C. (above the solidus).
5. The reaction with water vapour at 400° C. (below the solidus).
6. The effects of corrosion and surface films.

As a result of this study the authors conclude that the gas obtained from cast and rolled material consisted almost entirely of hydrogen. The diffusivity of hydrogen in the 7% magnesium alloy has been measured over a range of temperatures. The results can be expressed in the form :

$$D = D_0 e^{-ER/T},$$

where D_0=340 sq. cm./sec. and E=24,600 cal./g.-atom.

An approximate measurement of the solubility of hydrogen in the alloy at atmospheric pressure has been made at 500° C. It was found to be of the order of 0·05 c.c./100 g.

The reaction of the alloy with water vapour in the temperature range between the solidus and liquidus has been studied. The metal blisters and blackens characteristically, and a large quantity of hydrogen is formed, some of which enters the metal. Pre-exposure of the specimen to hydrochloric acid vapour had no effect, but pre-exposure to hydrofluoric acid vapour inhibited the reaction. It is concluded that hydrofluoric acid produces an impermeable film on the liquid phase.

The water-vapour reaction at 400° C. is less vigorous, since a protective film is formed. The amount of hydrogen absorbed by this specimen in this reaction would be about 0·7 c.c./100 g. for 18-gauge sheet after 2 hr. While this quantity would not cause blistering below the solidus in sound material, it would contribute to blistering if the metal were subsequently heated above the solidus, as—e.g., in welding.

The effects of annealing in damp air are essentially similar to those of annealing in water vapour.

By extrapolating from the hydrogen-diffusivity measurements, the diffusivity of hydrogen in the 7% magnesium alloy at room temperature is found to be about 2×10^{-16} sq. cm./sec. It follows that any pick-up of hydrogen through corrosion can only affect an extremely thin layer of metal—e.g., not more than about 0·00003 in. after one month at room temperature. Sheet with a corrosion film yields a considerable amount of hydrogen on heating *in vacuo*. The bulk of this gas is formed on heating by reaction between ions from the metal and water in the film. The amount released at 400° C. is comparable with the amounts obtained in steam exposures at this temperature. It is suggested that the corrosion film acts in much the same way as water vapour in the atmosphere. Thus, it contributes to blistering if the metal is heated above the solidus.

While surface reactions may contribute to blister formation below the solidus by increasing the gas content, it is improbable that they cause blistering unless severe flaws are present in the metal. Thus, an experimentally-produced flaw in overdrawn wire was enlarged considerably by annealing in moist air. It is suggested that another possible cause of blisters is the inclusion in the metal of particles of oxide with combined or absorbed water—e.g., in rolling or extrusion, which would react on heating to release hydrogen.

The Influence of Some Different Surface Preparations on the Corrosion of Magnesium-Base Alloys of Low and Normal Iron Content

By F. A. Fox, M.Sc., and C. J. Bushrod, M.Sc.

CORROSION tests, carried out on magnesium-aluminium alloys by immersing specimens in 3% solutions of sodium chloride, have shown that the method of preparing the surfaces of the specimens has a profound influence on their subsequent behaviour. Specimens prepared on 1 F emery paper corrode much more rapidly than similar specimens prepared with glass paper, pumice powder, or with successive grades of emery finishing with a polishing treatment, which all corrode at about the same rate.

It has also been found that, though specimens of varying iron content all corrode at substantially the same rate when prepared on I F emery paper, the other preparations result in metal of low iron content corroding at a slower rate than that of higher iron content.

Examination of emery removed from the I F paper showed that it contained about 7% of iron as Fe_2O_3, and it is suggested that the adverse effect of emery preparation is due to iron oxide pick-up by the metal surface. This suggestion is discussed in relation to the experimental results of the authors and other workers.

Equilibrium Relations and Some Properties of Magnesium - Lithium and Magnesium - Silver - Lithium Alloys

By W. Hume-Rothery, M.A., D.Sc., G. V. Raynor, M.A., D.Phil., and E. Butchers, M.A., D.Phil.

A SURVEY of the general properties of the elements suggests that the only useful alloys likely to have densities less than that of magnesium are those with a magnesium-lithium base ; beryllium is too rare, and sodium is too reactive and of too low a melting point to be of practical value as the major constituents of alloys. Lithium salts are not uncommon, and although

the pure metal is at the moment very rare, there appears no reason why it should not be prepared on a reasonably large scale.

Alloys consisting of the binary magnesium-lithium α-solid solution would not be expected to show very marked mechanical strength. It was thought possible, however, that alloys of an age-hardening or precipitation-hardening type might be prepared, consisting essentially of the magnesium-lithium base, with small additions of a third element, and for this purpose silver was suggested, because it possessed a suitable atomic diameter.

In the work described in this paper the α-solidus and α-solidus-solubility curves of the system magnesium-lithium have been determined accurately by the microscopic examination of alloys annealed to equilibrium. At high temperatures the solubility of lithium in solid magnesium diminishes slightly with increasing temperature; in this range of temperature the α-solid solution is in equilibrium with the β-phase, which possesses a random body-centred structure, and the equilibrium appears to resemble the α/β-brass type, in which the phases concerned are face-centred cubic and random body-centred cubic. At low temperatures the solubility of lithium in magnesium diminishes slightly, and there is some evidence that the β-phase develops a superlattice of the Fe_2Al type.

The constitution of magnesium-rich magnesium-silver-lithium alloys has been determined at 421°, 321° and 205° C. The isothermal sections show clearly that a wide range of alloys exist which will be homogeneous after solution treatment at 421° C., but two-phase or three-phase after further treatment at lower temperatures. It is interesting to note that as much as 13 at.-% of lithium can be introduced into magnesium-silver alloys without taking the resultant alloy outside the ($\alpha + Mg_3Ag$) field, so that alloys of the ($x + Mg_3Ag$) type can be obtained which are lighter than magnesium and do not contain a highly reactive constituent. It is also interesting to note that alloys in the homogeneous β-phase area, containing 26–30 at.-% of lithium, have a melting point of over 580° C. and a body-centred cubic structure, and consequently they offer the possibility of a ductile alloy considerably lighter than magnesium. From the purely structural point of view, therefore, these alloys appear of interest.

The question of corrosion-resistance has not been studied in detail, but a few preliminary tests have been made. Representative alloys containing up to 20 at.-% of lithium and a few atomic per cent. of silver have been exposed to an ordinary domestic atmosphere for five months, and do not show any obvious signs of corrosion to a greater degree than that of alloys of magnesium with more stable metals. There has been no tendency for the alloys to crumble to powder, as had been feared might be the case.

In general, the present work suggests that the preparation of a very light alloy with a magnesium-lithium base is not as impracticable as might have been feared.

Tin During the War

THE Hague office of the International Tin Research and Development Council has issued its first post-war Statistical Bulletin (dated February, 1946). This publication aims at bridging the gulf between the pre-war period and the peace period now ushered in. However, a number of sources are still inaccessible, or have even been entirely lost by the war, thus causing many gaps which it will be possible to fill only gradually. In many cases it was necessary to make an estimate. From the Bulletin we derive the following information:

Prices.—The tin prices in U.S.A. and England were fixed by the Governments during the war. The yearly average prices show the following picture:

	1939.	1940.	1941.	1942.	1943.	1944.	1945.	
U.S.A.	50·20	49·82	52·01	52·00	52·00	52·00	52·00	$ cts./lb.
England	226·3	256·6	261·7	275·0	275·0	300·0	300·0	£/long tons

Production.—World's tin production, which amounted to 149,700 tons in 1938, was stepped up in 1939 and 1940, reaching its peak of 245,500 tons in 1941 (estimated). The highest production before this date was recorded in 1937 (209,100 tons). As a result of the war in East Asia, this production moved on a much lower level after 1941.

The movement in world's production (these figures are partly estimated, and must be regarded as incomplete) during the war was as follows:

	1939.	1940.	1941.	1942.	1943.	1944.
Long tons	183,500	235,000	245,500	125,000	121,500	109,000

The export-quotas of the signatories to the International Tin Control Scheme amounted on an average to 74% in 1939, to 115% in 1940, to 130% in 1941, and were fixed at 105% as from January 1, 1942. The latter quota was based on the standard tonnages increased on the same date.

Consumption.—No estimate can be made yet in respect of world's tin consumption during the war years. The understated table shows the tin used in manufacture in the U.S.A. and England:

	U.S.A. Long Tons.		England. Long Tons.
1939	70,315		27,279*
1940	75,000		29,225*
1941	100,870		30,000†
1942	60,000		23,478
1943	56,000		17,631
1944	65,000		18,435
1945	33,000*†	(Jan.-July)	11,975 (9 months)

* Apparent consumption.
† Estimated.

Owing to destruction done to several of the large smelters, world's smelting capacity will presumably show a temporary decline, which is counterbalanced, however, by the fact that during the war an American-owned smelter of great capacity was established in the U.S.A. It may be assumed that also at the present moment the total smelting capacity available will be considerably in excess of world's production.

Purchase of Tin Concentrates in Malaya

PURCHASES of stocks of tin concentrates made in Malaya by the Tin Ore Buying Agency totalled 3,200 tons up to the end of January, 1946. Purchasing by the Agency is on a basic tin price of £300 per ton f.o.b. Malayan port. As from February 16, there has been a readjustment of the deductions made by the Agency against the producer; the treatment charge (hitherto paid by the Agency) is now included amongst the deductions, but the other Agency charges have been substantially reduced. The concentrates bought are to be smelted in Malaya.

MICROANALYSIS
CHEMICAL AND PHYSICAL METHODS
APPARATUS METALLURGICAL APPLICATIONS TECHNIQUE

IT can be argued with justice that there are several ways in which the merits of British microchemistry may be proclaimed. Firstly, there comes to mind research, and on this score no one can complain. Secondly, teaching offers an opportunity too often neglected. Prior to the War, this country almost completely, and largely unnecessarily looked to Germany as the fons et origo, not only of micro methods, which in part it was, but also of all authoritative teaching on the subject. This was false logic. Thirdly, and this is indeed a sore point, the excellence of British apparatus should be made manifest abroad, as much as at home. We are, at present, in the middle of an export drive. What, then, are we to think of a recent letter to a colleague from a prominent European microchemist. In this letter she complained that she wished, indeed, was extremely anxious, to restock her ravaged laboratory with British apparatus. Owing, however, to the unsurmountable difficulties placed in her way in every direction, she was regretfully forced to place the order in America. Whose fault this is we make no attempt to decide. The manufacturer, the Government, or just general muddle. In any case, this incident indicates little appreciation of the fact that things which appear minor to-day may have important repercussions on British chemistry in ten years' time. We must use all the ways. To rely only on research, as we would appear to be doing, is short-sighted loading of the dice against us that ultimately could work disaster.

The Potentiometric Micro-determination of pH: Part II

By J. T. Stock and M. A. Fill

The first part of this article described hydrogen and quinhydrone electrodes for small-scale work. Below, various types of glass electroder are discussed.

In 1909, it was discovered that the potential difference between a glass surface and a solution in contact with it depends on the pH of the solution.[18] From this discovery, a most important method of determining pH values has been developed. The arrangement is depicted schematically in Fig. 6. Within the thin-walled glass bulb, A, which dips into the solution to be tested, is an electrode-solution system of fixed potential, e.g.; a silver—silver chloride electrode immersed in 0·1 N hydrochloric acid. The cell is completed by connecting the test solution to a calomel—or other reference electrode, B, in the usual way. Unless the glass membrane is extremely thin, [19, 20, 21, 22] or its effective area is large,[23] the ordinary Poggendorf method of measuring the e.m.f. cannot be used. Instead, a valve electrometer, which draws negligible current from the system, has to be employed.[24] Suitable instruments are available commercially[25]. Since the glass electrode requires no hydrogen or other agent, allows measurements to be made very rapidly, and possesses certain other advantages,[1] the literature dealing with it is extensive.

Examples of the bulb- or Haber-type electrode as applied to the examination of small volumes of samples are the assemblies of Seekles[26] and of Lin.[27] The reentrant or cup type of glass electrode, shown diagrammatically in Fig. 7A, was introduced by Mrs. Kerridge[28] in 1925. It readily allows small quantities (of the order of 0·5ml.) of a sample to be examined, and is thus a pioneer among glass electrodes for making pH measurements on the micro scale.

In 1929, MacInnes and Dole[29] described an apparatus by which the pH of a single drop of material could be determined. The essentials are shown in Fig. 7B. The calomel electrode A is provided with a reservoir, B, filled with saturated potassium chloride solution. Side tube, C, closed by a short length of rubber tubing, which may be compressed by a screw clamp, allows the level of potassium chloride solution in the capillary outlet, D, to be adjusted. The tip of D is ground flat and serves to receive the sample, which is prevented from spreading by lightly waxing the outside of the

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25 See, for example, List 108A, Cambridge Instrument Co. Ltd., London.
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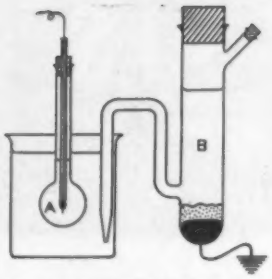

Fig. 6—Schematic arrangement of Glass Electrode System.

tip. To operate the apparatus, the screw clamp is tightened until the capillary outlet is swept out with potassium chloride solution, the excess being wiped from D with a piece of filter paper. A drop of sample is placed on D, and the clamp loosened slightly in order to suck the liquid back into the capillary for about 1 to 2 mm. After again wiping the flat tip of D, a second drop of the sample is placed upon it. The shield F is then placed in position. A mechanical device allows the membrane type glass electrode E,[30] the lower vertical surface of which is lightly coated with paraffin, to be cautiously lowered until the thin membrane makes contact with the drop of sample, when the e.m.f. is quickly measured.

A similar apparatus was later described by Duspiva.[31]

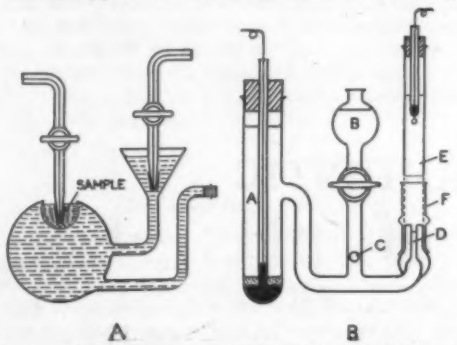

Fig. 7.—Glass Electrode Systems for working with small quantities of sample. A—Kerridge. B—McInnes and Dole.

In his apparatus, the sample is placed upon the membrane of the glass electrode, while contact with the reference electrode is made by lowering a capillary filled with agar-potassium chloride gel until the tip touches the test drop.

Of other electrode assemblies of the diaphragm type which are suitable for dealing with small quantities of samples, mention may be made of those described by DuBois,[32] Païc,[33] de Barenne and his co-workers,[34] and by Stadie and his co-workers.[35] The apparatus designed by the latter team is available commercially in this country,[25] the principal modification being that a miniature bulb-type electrode is used in place of the diaphragm type originally described. The electrode chamber has a capacity of only about 0·2 ml., and is surrounded by a water jacket to stabilise the temperature.

Glass electrode assemblies in which the membrane is in the form of a capillary have attracted no little attention.[23, 36, 37, 38, 39, 40] The capillary technique permits the examination of samples having a capacity of only 0·005 of a millilitre, as was shown by Claff and Swenson.[40] The principle of their apparatus is shown in Fig. 8. Capillaries of Corning 015 glass, which may be extremely fine, are used. Insulating discs are attached by picein, so that the capillary may be suitably supported in a simple jig. The fan-shaped funnel is filled to capacity with 0·1 N hydrochloric acid, in which dips a silver—silver chloride electrode. After having introduced the sample into the capillary, the latter is supported so that the middle portion is immersed in the solution in the funnel. The assembly is then raised until junction is made with the specially designed saturated calomel electrode, when the e.m.f. of the cell is determined.

In this short review, it is hardly possible to discuss more than briefly some of the electrode systems which have been employed. Nevertheless, it is hoped that the general tendencies have been indicated.

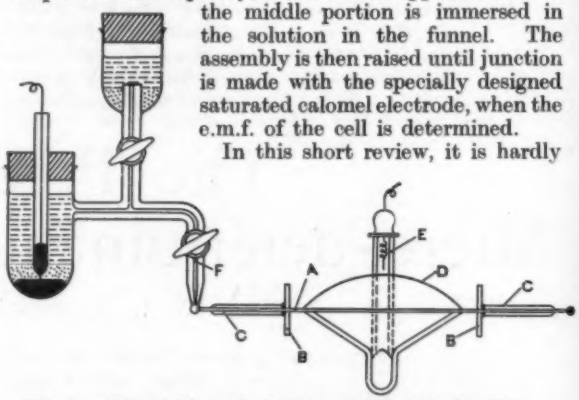

Fig. 8.—Glass Electrode System employing Capillary Form of Membrane. A—Capillary containing sample. B—Insulating discs. C—Paraffin coating. D—Fan-shaped funnel. E—Silver—Silver Chloride Electrode. F—Calomel Electrode.

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Miscellaneous Microchemical Devices

By J. T. Stock and M. A. Fill

1. An Automatic Micro-Pipette for Spot and Drop-Tests

DURING a recent investigation, an attempt was made to carry out various drop reactions under the microscope in such a manner that the observations could be rendered at least semi-quantitative. In order to render uniform the drop sizes of test solution or of reagents, several methods were used until a simple automatic pipette was developed. Several of these were constructed and used throughout the work with complete satisfaction.

The construction of the device is shown in Fig. 1. The outer portion is made from tubing of 3 mm. bore, which is drawn down and cut off to form a delivery end having a bore of approximately 1·5 mm. at the tip. Vent hole

A prevents over-filling of the pipette, as well as allowing ingress of air. Constriction B at the upper end permits axial movement of plunger C without side-shake. The lower end of the plunger is drawn down, lightly ground into the delivery end and cut to length, so that it projects about 0·5 mm., as shown at D.

By means of a medicine-dropper or a piece of drawn-down tubing, the solution or reagent is introduced through the vent hole. Holding the pipette as nearly vertical as possible between the index finger and thumb, and with the fore-finger resting lightly upon the plunger head, the pipette is operated by bringing down the tip sharply and squarely on to the slide, test plate, or paper. This causes the projecting end of the plunger to be forced upwards until it is flush with the tip of the delivery end, thus permitting flow of liquid.

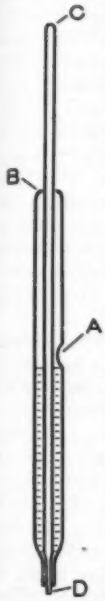

Fig. 1.—Automatic micro-pipette.

Delivery is stopped by pressing smartly upon the plunger head; this causes the tip of the outer portion to be lifted clear by the natural reaction of the index finger and thumb. The complete operation of delivery is extremely rapid and should not exceed one second. Once the technique has been acquired, reproducible delivery is a simple matter. This is demonstrated by the results in Table I, which were obtained by delivering 10 successive drops on to a tared microscope slide and then weighing rapidly. The slide was then cleaned and a further series of drops deposited.

Provided that the change in level of the liquid within the pipette is not too great, it has but a small effect upon the amount delivered. This is shown by the results in Table II, which were obtained by delivering six successive series, each of 10 drops without refilling the pipette.

TABLE I.—DELIVERY ON TO MICROSCOPE SLIDE.

Bore of tip .. .. .. 1·35 mm.
Dia. of plunger tip .. 0·90 ,,
Projection .. .. .. 0·67 ,,

Series.	*Average weight per drop, mg.*	*Difference from mean, mg.*
1	0·75	0·00
2	0·75	0·00
3	0·77	+0·02
4	0·68	−0·07
5	0·76	+0·01
6	0·77	+0·02

Experiments upon the spotting of filter paper were also carried out. Pieces of Whatman No. 1 paper were cut to 3 in. × 1 in., weighed, supported upon a clean microscope slide, and re-weighed after spotting with 10 successive drops. The spots were quite symmetrical and of reasonably uniform size as is shown by the results in Table III.

TABLE II.—DELIVERY ON TO SLIDE WITHOUT REFILLING.

Bore of tip .. .. .. 1·35 mm.
Dia. of plunger tip .. 0·90 ,,
Projection .. .. .. 0·45 ,,

Series.	*Average weight per drop, mg.*	*Difference from mean, mg.*
1	0·31	+0·01
2	0·29	−0·01
3	0·29	−0·01
4	0·30	0·00
5	0·31	+0·01
6	0·29	−0·01

Considering the simple nature of the device, the results are fairly satisfactory. There is no doubt, however, that *closely* reproducible results are not readily obtainable,

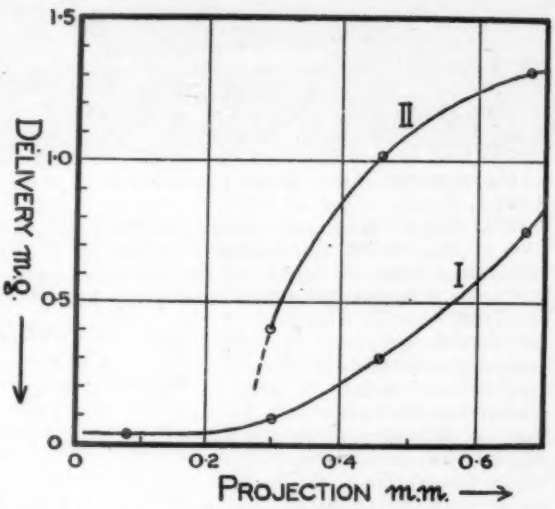

Fig. 2.—Effect of plunger projection upon delivery of solution.

TABLE III.—DELIVERY ON TO FILTER PAPER.

Bore .. .. 1·35 mm.
Dia. of plunger tip .. 0·90 ,,
Projection .. .. .. 0·67 ,,

Series.	*Average weight per spot, mg.*	*Average dia. of spot, mm.*
1	1·31	6·1
2	1·33	6·5
3	1·20	5·8
4	1·34	6·3
5	1·31	6·0
6	1·26	5·8

owing mainly to the enhanced dependence upon the period of delivery and to the varying penetration of the fine tip of the plunger. As is clear on comparing the results in tables I and III, obtained by using the same pipette, the delivery per operation on to filter paper far exceeds that on to glass, although identical timing and technique were used for both sets of experiments.

Early on in the investigation, it became clear that the projection of the lower end of the plunger was a major factor in controlling the delivery. Accordingly, a pipette having essential dimensions similar to those quoted in Table III was prepared. The projection of the lower end of the plunger was first measured by means of a microscope having a micrometer eyepiece. The average deliveries, both on to glass and on to filter paper, were then determined as described above. After shortening by very careful grinding, the projection of the lower end of the plunger and the average deliveries were again determined. After further shortening, the measurements were repeated. The results are shown in Fig. 2. With a projection of less than 0·3 mm., delivery on to filter paper was found to be too variable to be satisfactory; when the projection was decreased to 0·1 mm., no delivery on to filter paper was obtained. Quite otherwise were the results when delivering on to a microscope slide. Satisfactory operation was obtained even with a projection of less than 0·1 mm., when the average delivery per operation was found to be 0·047 ± 0·003 mg.

We have also used the device for delivering indicators in microtitrations, either directly in to the (dry) titration cup before introducing the solution to be titrated, or by depositing it on to the flattened end of a micro-stirring rod and then thrusting this into the solution.

II. A Simple Method for Agitating Liquid Baths

NO moving parts are involved in this method of agitation, which is operated by water-pump suction It is applicable to any mobile, relatively non-volatile liquid, either at low or at elevated temperatures. Two typical uses are the agitation of a small melting-point bath containing heated liquid paraffin and of a deep water-bath maintained at 10° C.

In its simplest form, the agitator consists merely of a wide T-piece which is adapted as shown in Fig. 1. Dimensions naturally depend on the particular application ; for a melting-point bath containing 100–400 ml of liquid, tube of about 1 cm. bore is suitable. The extremity A of the bent side arm is constricted to a diameter which is about half that of the tube, and is supported to dip a few mm. below the surface of the liquid when the latter is at rest. The portion B of the tube reaches almost to the bottom of the bath.

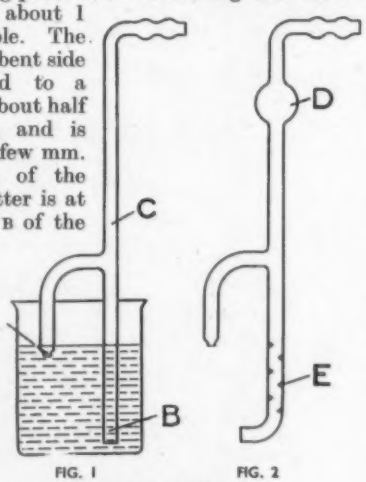

FIG. 1 FIG. 2

Fig. 1.—Agitating device.
Fig. 2.—Modification to prevent frothing.

On starting the pump, the liquid is drawn in through both inlets A and B, and rises in C. The level in the bath thus falls a few mm. until ingress of air through A momentarily destroys the suction, causing the liquid in C to fall sharply. It mixes with that ejected from the side arm and flows out at B. The level of the bath thus being restored, the cycle is repeated. The total effect is the continued transference of liquid from the top to the bottom of the bath, accompanied by slight up-and-down oscillation of the surface. The degree of agitation may be controlled by adjusting the suction.

Since the operation of the device depends upon the depth of immersion of inlet A with respect to the static level of the bath-liquid, rapid evaporation is undesirable. Hence for prolonged heating at elevated temperatures, liquids of high boiling-point are preferable. When frothing is encountered, there is some tendency for the liquid to be carried over into the pump. To prevent this, a bulb may be blown on the upper part of the tube as shown at D, Fig. 2). To enhance the mixing effect, indentations E may also be added to promote turbulence.

III. A Device for Drop Reactions with Small Quantities of Solutions

DURING a study of reactions in capillaries[1] the device described below was evolved to investigate the operation of mixing. Besides permitting tests to be carried out with small quantities of solutions, transference rods or capillaries cannot become contaminated, except with a trace of distilled water, and mixing is performed without the use of stirring rods or other auxiliaries.

Glass rod of diameter approximately 2 mm. is drawn down into a thread about 0·2 mm. in diameter and about 15 cm. long. Some 10 cm. of the thread is cut off and used for making the parallel-sided loop, which has an opening 30 mm. long and 0·5 mm. wide. To do this, the thread is carefully heated in the middle and bent into hairpin form. The free ends are fused together, and joined to the handle, as shown at (*a*) in Fig. 1. The loop is then bent into a " V " having an angle of approximately 120°, as shown at (*b*). The hook at the end of the handle allows the device to be hung on the rim of a beaker, thus facilitating safe storage and cleaning.

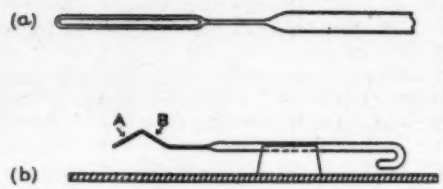

Fig. 1.—Device for drop reactions.

Fig. 2.—Modification for use with two or more Reagents.

When required for use, the loop is dipped in distilled water, and hung downwards for a few seconds to permit drainage. The droplet which collects at the bottom is removed by touching with a pointed strip of filter paper. The device is then placed on a support made by grooving a cork, the angle of the " V " being uppermost, as at (*b*). A small quantity of the test solution is now introduced at point A. This may be performed by dipping a fine glass rod into the solution and then drawing the rod cross-wise at *A* until a droplet or film of liquid forms.

In a similar manner, a droplet of reagent solution is formed on the other limb of the " V " at point B. The stem of the device is now rotated in the support until the angle of the " V " points downward. The droplets on the arms run into the angle and coalesce, when any reaction may be observed by holding over a white tile. Effective mixing or agitation may be performed by tilting up and down, so that the droplet flows from one end of the loop to the other. It was found that colour-reactions could be carried out with as little as 0·1 mg. of test solution.

In certain tests, e.g., confirmation of the presence of copper by production and destruction of the cuprammonium ion,[2] or the " Magneson " reaction of magnesium[2], [3], it is necessary to employ more than one reagent. The modification shown in Fig. 2 permits this to be done with ease. Instead of being bent to form a simple " V," the loop is corrugated by careful working in a micro-flame. After wetting and draining, a droplet of the test solution is introduced at A, while a droplet of the first reagent is deposited at B. On allowing the loop to hang downward, the drops coalesce. A droplet of the second reagent may then be introduced at C and allowed to flow into the mixture, while a third reagent may be applied at D. Provided that the loop is not tilted upwards, no contamination of the unused portions of the corrugation occurs, so that the transference rods may be returned to their respective solutions.

It is obvious that if it is desired to add several reagents simultaneously instead of successively, then the various droplets may be introduced at A, B, C, etc., and then caused to mix by tilting the loop.

1 See, for example, A. A. Benedetti-Pichler, " Microtechnique of Inorganic Analysis," Wiley, New York, 1942, p. 94.

2 C. L. Wilson, " An Introduction to Microchemical Methods," Methuen, London, 1938, p. 85.

3 Hopkin and Williams, Ltd., " Organic Reagents for Metals," 4th Edn., London, 1943, p. 103.

Micro-Hardness Testing of Materials

By V. E. Lysaght

THE Knoop indenter extended the field of indentation hardness testing to permit the study of surface conditions less than 0·005 in. in thickness and to test brittle materials, such as glass, porcelain, carbides, etc., which ordinarily shatter under testing loads. Surface conditions less than 0·005 in. in thickness could not be effectively studied even with the Rockwell superficial hardness tester or the Vickers tester. Advantages gained by using the Knoop indenter were of little value to the engineer or metallurgist unless it were possible to grind this indenter at a reasonable cost to meet the exacting specifications of the National Bureau of Standards. Recently a technique developed for producing Knoop indenters on a commercial basis, has produced the Tukon tester.

The Tukon tester can be used for the determination of Knoop hardness numbers under testing loads of 25–3,600 grams, and is fully automatic under electronic control in a synchronous cycle. The indentation is made by elevating the specimen against the indenter until the hardness of the specimen resists further indentation. As this condition is reached, an electronically controlled circuit prevents further elevation of the specimen and the load is applied, controlled by means of a timing motor, for about 20 seconds. At the end of this time interval, the indenter is automatically removed from the specimen and the elevating screw lowered away. Early in the development of this tester it was observed that, when testing very light loads (under 200 grams), the indentation was enlarged mainly due to the kinetic load of contact between indenter and specimen, and provision had to be made to prevent such an error creeping into the test.

In the hardness testing of metal surfaces the use of the Knoop indenter and loads of 50–2,000 grams have made possible the study of the hardness of the nitrided case of nitrided high-speed steels with depth of indentation marks less than one micron (0·00004 in.). This indenter has also shown that hard spots encountered when milling high-speed steels were primarily the result of tremendous heat locally developed by a failed tool. Hardness surveys have been made with the Knoop indenter to study changes in the hardness of the surface of hard steel caused by improper grinding, and to control the hardness of chromium plate less than 0·0005 in. in thickness. Micro-hardness measurements have also been of value in welding research for studying changes in the hardness of heat-affected zones.

In the manufacture of precision parts such as watch and clock parts, where hardness is an important physical property, micro-hardness testing with the Knoop indenter and Tukon tester has been developed and many such parts have been subjected to such testing both for experimental purposes and also for routine testing. Alloy materials tested have included phosphor bronze, copper, invar, nickel silver, beryllium copper, stainless steel, corundum, nickel and chromium plating and diffused indium plating. In the surface processing of metals the Knoop indenter has been used to study the hardening effects of the grinding treatment on steel plates having different metallographic structures and for the study of indentation hardness of the surface layer of cold-rolled steel plate and the relation of surface hardness to the mode of initial finishing. Tests made with the Tukon tester have shown the degree of surface hardening produced by present-day metallographic polishing techniques, Table I.

TABLE I.—APPROXIMATE DEPTH OF WORK-HARDENING IN MILS (0·001 IN.)

Polishing Method	Material and Rockwell Hardness			
	Copper E 29	Brass B 82	24's Aluminium B 61	Tempered Tool Steel C 35
Coarse grinding ..	5	1	1	0·2
Fine grinding ..	2	0·5	0·5	0·1
Metallographic papers	1·6	0·2	0·2	0·06
No. 1 wax wheel ..	1	0·1	0·1	—

Original studies on micro-hardness testing assumed that the elastic recovery of the long diagonal of the indentation could be considered negligible and that any increase in hardness at low loads was due to work-hardening of the surface. Recent work, however, carried out on a limited number of tests, has shown that this was not necessarily true, because on removal of the indenting load, the long diagonal of the impression could be observed to shorten and, if recovery were a constant amount, an apparent increase in hardness in low loads would result. It was, therefore, recommended that, especially with light loads, the indenting load be specified. Further work on this subject, however, appears to be necessary.

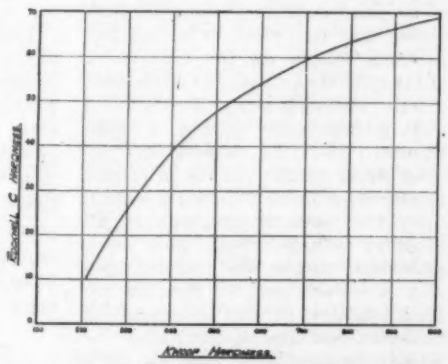

Fig. 1.—Relationship between Rockwell C-scale numbers and Knoop numbers. Load 500 grams.

From *Materials and Methods*, 1945, vol. 22, No. 4, pp. 1079-1084.

In the hardening of steel, micro-hardness determinations have been useful in induction hardening tests and in studying austenitic transformations where micro-hardness values of from about 190 for soft pearlite to about 720 for fully hardened martensite were obtained in an alloy steel containing 1·14% carbon.

A conversion chart has also been determined, showing the relation between Rockwell C scale and Knoop number using a load of 500 grams, Fig. 1, but such a chart should not be used to set up specifications or to meet specifications, nor should it be considered for Knoop numbers obtained for loads other than 500 grams.

In using the Knoop indenter on the Tukon tester samples must be properly prepared and must be mounted or supported so that the indenter is normal to the testing surface. Results obtained with both instruments must be interpreted with care, particularly where light loads are applied. Indentations less than 50 microns must be measured by experienced operators using optical equipment with the greatest magnification and giving the best defined outline of the indentation. Allowing for such limitations, micro-hardness testing is invaluable for measuring hardnesses of thin superficially hard materials, small parts and edges of cutting tools, hardness surveys at close spacing and for decarburisation studies.

Pumping Molten Magnesium

By M. M. Moyle

EXAMINATION of recent developments reveals to the metal-working industries the ease of handling magnesium in the molten state. One of the most striking of these developments, and one which is having far-reaching effects on the magnesium industry, is the pumping of the molten metal. Pumping is the simplest and most advantageous method of liquid transfer, and the adaptability of magnesium to the process is unique among the common structural metals. Since the metal is adaptable to all kinds of casting—ingot, billet, sand, permanent mould, and die-casting—and is widely used in these forms, improvements in the handling of molten metal are significant to the whole industry, and pumping is a major factor in the successful application of mass production techniques to magnesium.

The two properties of magnesium which contribute most to its ability to be pumped are its relatively low melting point, and its inertness to iron and steel at the operating temperatures. Because of the high temperature of molten iron, steel, brass and bronze, and because of the attack on iron by aluminium, these metals must normally be handled in special refractory equipment. In the case of zinc and lead the application of pumping is limited. At the present time the commercial use of pumps for molten magnesium is established in alloying processes, alloy ingot casting, continuous billet casting, and wherever bulk metal transfers are required. Molten magnesium is being pumped for distances varying from only a few inches to 25 ft., and in capacities ranging from a trickle to a ton per minute.

The adoption of the reverberatory melting furnace made possible the melting of large amounts of magnesium, and pumping has made large scale alloy production a continuous operation, since from the time the operator charges solid metal into the initial melting unit until another operator takes away a cooled ingot or billet, the metal is stirred and transferred entirely by pumps. In ingot casting, a large-capacity pump can be made to transfer metal from a reverberatory furnace to a large holding pot from which the metal is pumped continuously to ingot moulds. Stirring of the metal by means of the pumps ensures good mixing of alloying elements, and assists flux-refining operations, and at the same time eliminates the laborious method of hand stirring. With certain modifications of the operation in the ingot-casting process, pumps are also used for metal transfer throughout the continuous billet-casting process.

As the temperature of operation is relatively low, since magnesium alloys melt at approximately 650° C., ordinary cast-iron and steel pumps are suitable, and heat-resisting steels containing manganese, tungsten, chromium, vanadium or molybdenum, can be used to advantage when greater strength or oxidation resistance is desired. Steels containing nickel and cobalt as major alloying elements are not suitable in contact with liquid magnesium, as these elements are soluble in magnesium, and contamination of the melt and deterioration of the steel can be expected. The centrifugal pump with proper modification has been found the most suitable type of pump for use with molten magnesium. Piston pumps have been used successfully, but centrifugal pumps are easier to control, have higher capacity, and require a minimum outlay of equipment. Several centrifugal pump designs are in use depending on their specific application.

From *Metals and Alloys*, 1945, vol. 22, No. 3, pp. 716-720.

A White High-Manganese Brass

By R. S. Dean, J. R. Long, T. R. Graham, and C. W. Matthews

IN a previous paper[1], two of the authors reported on the mechanical properties of a 65% copper, 10% manganese, 25% zinc alloy as compared with similarly processed cartridge brass. Additional investigations on high copper alloys demonstrated that at lower zinc concentrations even larger amounts of pure manganese may be added to copper-zinc alloys without exceeding the single-phase solid solution limits, and that similar increases in tensile strength, yield strength and hardness may be obtained without excessive decrease in ductility. Several compositions in this range showed also a decided colour change with increasing manganese, the normal "brassy" colour being considerably lightened and reduced to a faint yellow cast at 15% manganese. With higher manganese contents the alloys were essentially white and similar in appearance to the nickel-silver alloys.

From this work it appeared desirable to determine more completely the properties of the wrought annealed white alloys, and accordingly, a more detailed study was planned of an alloy having the composition 70% copper, 20% manganese and 10% zinc. This alloy remained single-phase throughout the working and

From *Metals Technology*, 1945, Vol. 12, No. 4, and *A.I.M.M.E. Technical Publication No.* 1813, pp. 1-9.
[1] *Metallurgia*, 1944, Vol. 31, No. 182, pp. 111-112.

TABLE I.—COMPOSITION OF ALLOYS

No.	Cu	Mn	Zn	Al	Fe	Si
1	70·0	20·4	9·3	0·05	0·019	Nil.
2	68·6	20·1	10·5	0·17	0·420	0·11
3	74·5	19·8	5·1	0·13	0·017	Nil.
4	65·5	19·9	14·8	0·02	0·005	Nil.
5	69·4	15·9	14·5	0·08	0·021	Nil.
6	74·9	15·5	9·8	0·04	0·016	Nil.
7	64·7	25·6	9·4	0·04	0·029	Nil.
8	70·3	24·5	4·8	0·04	0·017	Nil.

annealing cycles of the preliminary tests and appeared to be about the strongest of the white single-phase alloys. The project included also the investigation of additional alloys deviating from this composition by 5% of each of the three components to evaluate the effects of variation in composition; also a study of the central alloy made with silicothermic manganese. The compositions of the various alloys are given in Table I.

Electrolytic manganese, wirebar copper, and electrolytic zinc were used for the preparation of the alloys. Representative analysis of the electrolytic manganese seldom exceeded 0·01% iron, 0·02% sulphur and 0·02% lead with no silicon or aluminium, while a typical sample of the silicothermic manganese contained 2% iron, 0·81% silicon, 0·16% carbon and 0·36% aluminium. Melting and alloying were carried out in a high-frequency furnace with magnesia crucibles. All ingots were hot-rolled at 650°–700° C. and then cold-rolled with intermediate annealing to produce sheet finished to 0·0625 in. thickness. Testing the central alloy after annealing for 1 hour at 650° C. after 60% cold reduction produced an average grain size of 0·065 mm. which was considered satisfactory for subsequent working and was adopted for the annealing cycles. The silicothermic manganese alloy consistently had a 0·01 mm. grain size. This small grain size is characteristic of alloys made with silicothermic manganese and in subsequent annealings at higher temperatures it continued to show a much smaller grain size than the electrolytic manganese alloy at any given temperature.

The cold rolling and annealing tests carried out show that when the alloy is made with electrolytic manganese it has in the annealed condition (0·065 mm. grain size) a tensile strength of 26 tons per sq. in., a yield strength of 8·0 tons per sq. in. with an elongation of 46·5% and a harden of 44 Rockwell B. If the alloy is made with silicothermic manganese, it is stronger, harder and less ductile than made with electrolytic manganese, as the normal impurities in silicothermic manganese produce an ageing effect in 200°–315° C. range, raise the recrystallisation temperature, and restrict the normal grain growth in the 540°–650° C. temperature range. The properties of alloys deviating from the central alloy by 5% for each of the three components show the alloy to be insensitive to relatively large variations in composition. The alloy is also slightly stronger and harder than straight alpha brasses and compares favourably in mechanical properties with nickel silver and other common high-copper alloys.

Rhenium Reported in North Carolina Molybdenite

THE U.S. Bureau of Mines has found that the molybdenite in Halifax County, North Carolina, carries rhenium in concentrations well above the amount previously known. The analytical reports show rhenium values of 50, 80 and 90 parts per million (or 150 calculated for pure molybdenite), compared to the normal of 10 parts per million. Since previous determination of rhenium give values ranging from 1·2–65 p.p.m., with 597 p.p.m. as the highest for a few fissure specimens, and 7·46 p.p.m. for a molybdenite concentrate, these concentrations can be regarded as unusually high. The spectrographic examination left little doubt of the presence of large quantities of rhenium.

Rhenium is a rare element of the manganese family. The existence of two members of the manganese family, eka and dvi manganese, was long suspected, and Mendeljeeff left places (43 and 75) for them in the orderly table of periodic system. In 1925 Noddack and Tacke made an effort at the concentration of the element dvi-manganese from platinum ores and from columbites, and definite evidence was announced by Berge and Tacke for its presence in a chemical concentrate. The name rhenium was assigned to it. However, the results of these chemists were declared inconclusive, and discovery of element 75 was claimed by J. G. F. Druce and his co-workers, U. Dolejsek and J. Heyrovsky, who found rhenium obtainable as $KReO_4$ (potassium perrhenate), a colourless stable salt, analogous with potassium permanganate. Prandtl was unable to find element 75 in the manganese salts.

Rhenium has useful properties as a catalyst and for electrical filaments. It may be alloyed with rare metals, and can be deposited electrolytically. Its resistance to hydrochloric acid is an important and unusual property. An impressive list of patents indicates that the metal and its compounds may become commercially important. Some proposed uses are for electric heating devices, incandescent lamps, radio and power tubes, pen nibs, and in the manufacture of chloroform and production of anti-knock fuels.

From *Engineering and Mining Journal*, April, 1945.

Vanadium

THE element vanadium was discovered in 1801 by Andres Manuel del Rio in a sample of lead ore from Zimapan, Mexico. Thirty years later the chemist Nils Gabriel Sefstrom, of Stockholm, recognised the new element in a sample of titaniferous magnetite from Taberg, Sweden. In 1867 Sir Henry E. Roscoe and his assistant, Sir Edward Thorpe reduced vanadyl trichloride $BOCl_3$ to a metallic mass possessing a bright silver-white lustre. Vanadium continued to be a rare metal until 1905, when Riza Patron discovered the extensive deposits of vanadium sulphide in the Andes of Peru.

Vanadium is widely distributed in the earth's crust, but commercial deposits are rare. The amount of vanadium in the crust is equal to the combined weight of copper, zinc and lead. An unusual occurrence of this element is in certain coals and oils. Italy has for years derived the metal from this source. Flue dust from the boilers of ships burning Venezuelan fuel oil contains from 20–40% V_2O_3. In the United States vanadium is being mined in Arizona, Colorado and Utah. In 1940 deposits from these areas yielded about 3,000 tons of vanadium, approximately half that year's consumption of the metal. Important commercial minerals are patronite, carnotite, roscoelite and vanadinite.

Pure, ductile vanadium was first prepared by J. W. Marden and M. N. Rich in 1927. A mixture of calcium chloride, calcium metal and the oxide V_2O_3 was heated in a steel bomb at 950° C., and the metal obtained analysed 99·3–100% vanadium. Another method of obtaining relatively pure

From *Bull. Electrochem. Soc.*, Dec., 1945.

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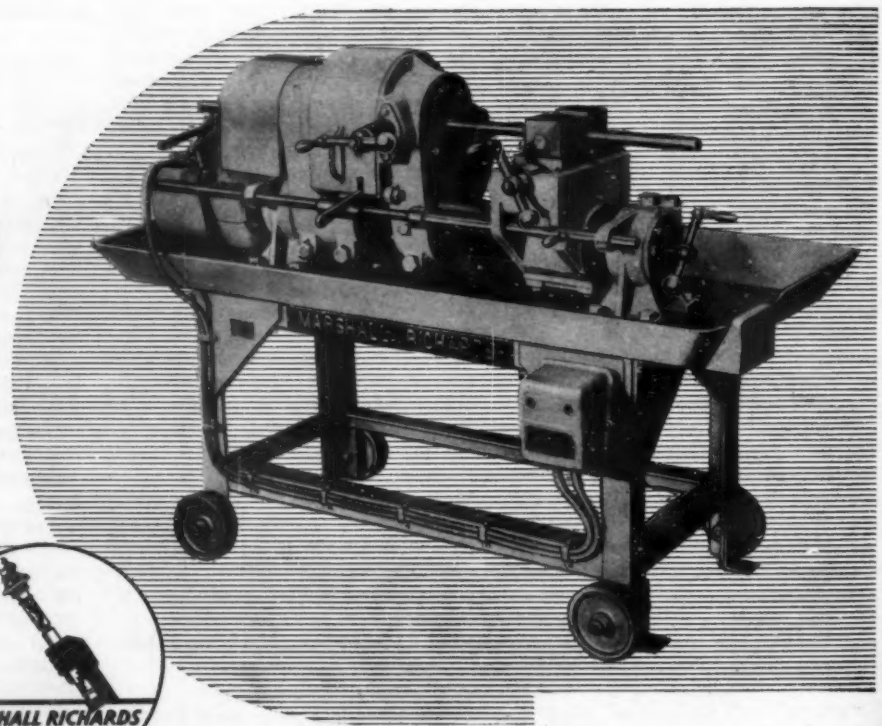

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vanadium is by the thermal decomposition (de Lodyguine) of the diiodide VI_2.

The commercially important application of vanadium is in steels, in particular high-speed steels (0·5–1·0% V). Vanadium steels are highly resistant to strain, shock, and vibration. For the steel industry vanadium is furnished as ferro-vanadium, containing about 40% V, which melts at about 1,480° C. In the oxide form, V_2O_3, vanadium is a widely used catalyst (contact sulphuric acid, etc.).

Vanadium metal crystals are body-centred cubes. The metal is bright, silvery white and ductile. Its melting point is 1,715° C.; its specific gravity 6·0. There is only one atomic species (no isotopes). The work function is 3·79 e.v. It is soluble in nitric acid but resists attack by hydrochloric acid. Pure vanadium is stable in air. At red heat it forms a nitride. The valences are 2, 3, 4 and 5. The monoxide, VO, behaves like an element and forms vanadyl compounds. Vanadium heated to 1,300° C. in hydrogen forms a hydride.

Relation Between Life Testing and Conventional Tests

By R. E. Paterson

WHEN a new engine, a new electric motor or a new domestic appliance has passed a satisfactory shop test, it is sometimes necessary to repeat the performance of the particular unit, starting, stopping and cycling, year in and year out for a satisfactory life for the equipment. To obtain such data life tests are devised to test component parts usually with the idea of crowding a life of cycling into a reasonably short testing time. The fatigue testing of an aeroplane generator bracket, or of a turbine blade at high temperature, the testing of appliance plugs and terminal pins which must function satisfactorily for 30,000 operations, and a device for the fatigue and wear testing of the opening or closing of a refrigerator door, which operates at 15 cycles per min. and which covers an equivalent life of 15 years in 12 days, are examples of such testing. Most life tests, which should duplicate service conditions as closely as possible, are rather complicated and usually do not even resemble the tests made in conventional material testing laboratories.

In analysing life testing in terms of its fundamental elements, time and cycling are the important characteristics and lead to consideration of certain conventional tests such as fatigue, creep (including relaxation and creep rupture), wear (including erosion and fretting), corrosion (including deterioration of non-metallic materials and lubricants), and repeated impact. These tests may be divided into two groups: Group A—quantitative tests, capable of giving specific design data, and group B—qualitative tests, useful primarily for a comparison of materials. If the main problem of a new machine is in the second group, say a wear problem, it cannot be solved by looking up data and calculating the answer, nor can the desired design data be obtained by making a conventional test. A life test simulating actual conditions as closely as possible, is very much in order for problems of this kind. If the problem is in the first category, however, that is fatigue or creep, there is at least a possibility of predicting satisfactory performance, but even then there are more considerations involved than is generally appreciated.

Consider fatigue in bending in the fillet section of an axle made of normalised S.A.E. 1035 steel (0·32–0·38% carbon) with a nominal endurance limit of 15 tons per sq. in. Such a fillet, $\frac{1}{16}$ in. radius, causes a local increase in stress expressed in terms of a stress concentration factor, K_t = maximum stress/nominal stress. In a shaft fillet or groove there is in addition to the stress in the bending plane a stress at right angles in the circumferential direction, and these two stresses are related by a constant "theoretical strength reduction factor" K_t, which for the shaft under consideration is equal to 1·85. High-strength heat-treated steels are highly sensitive to notch effects, while normalised steels are less sensitive, and such sensitivity is usually expressed by a "sensitivity index," q, where

$$q = \frac{K_f - 1}{K_t - 1}$$

and

$$K_f = \frac{Se}{S_n} =$$

Endurance limit without Fillet

Nominal endurance limit with Fillet.

From this definition q is defined in such a way that for no notch effect in fatigue (that is $K_f = 1$) q = o, and for full theoretical notch effect (that is $K_f = K_t$) q = 1, and this gives a scale from 0–1 for evaluating notch sensitivity. From this scale, q = 0·82 for the shaft under consideration.

The maximum nominal bending stress, therefore, for a shaft of S.A.E. 1035 steel, having $r = \frac{1}{16}$, and q = 0·82, is 9 tons per sq. in., and if such a shaft is a simple machine element with operating stresses substantially lower, it should not be necessary to make life tests.

In considering life tests for units involving creep, data are of interest in steels primarily at elevated temperatures, but for die castings and other soft metals important effects occur at room temperatures. In the case of relaxation, integrated effects of the behaviour of large masses of grains are involved and therefore analytical methods are appropriate as a means of interpreting and extending test data. To predict creep in the red hot part of a turbine in 10 or 20 years, it is necessary to extrapolate by mathematical means from the results of creep test of a few months' duration. In the case of creep rupture (brittle failure after an extended time at constant load and temperature), the experimental approach is best.

In general, life testing is applicable chiefly to items of large production. For certain problems involving wear friction, lubrication, etc., it is the only satisfactory test and for many complicated fatigue problems a life test is most certainly advisable.

From *Bulletin American Society for Testing Materials*, 1945, No. 133, pp. 9–16).

Protection of Heated Vessels against Acid Corrosion

By K. Frank

HEATED vessels which are required to contain acids, and are of large dimensions are generally constructed of steel, cast steel or cast iron, being lined or clad with an acid-resisting metal, or alternatively enamelled or lined with rubber, synthetic resin or a ceramic protective layer. The metallic protective coatings generally require a considerable use of metals in critical supply; the thicker, non-metallic protective layers have the disadvantage of a smaller heat transfer factor. Coatings of enamel are sensitive to impact and cannot be repaired *in situ*, if they become damaged. The disadvantages mentioned are avoided by means of a new lining process, by which on the inside wall of the steel or iron vessel, are soldered ferro-silicon plates containing about 70% silicon. These plates are cast in chill

Chem. Technik, vol. 15, 1942, pp. 235-240.

moulds, and subsequently annealed. The plates are made somewhat broader at the foot, and, as solder, a tin or lead-cadmium solder is employed. The joints between the plates are closed with a dense and acid-resistant organic or inorganic cement. As the soldered layer does not damn the heat transfer effect to any practical extent, heat transfer figures are reached amounting to about three-quarters of the vessel in the unlined condition.

Non-Axial Loading of Grey-Iron Tensile Specimens

By J. O. Draffin and W. L. Collins

IT is generally agreed that with material such as grey cast iron, the non-axial loading of tensile test specimens results in a lower strength than would be obtained by axial loading. The extent of this reduction in strength, the conditions which will produce it, and how these conditions may be remedied are questions of importance to those whose product is the subject of such tests. A limited number of tests has therefore been made under two conditions which will produce non-axial eccentric loading. These conditions were: (1) that the specimen holders are not free to adjust themselves under load, and (2) that the central portion is not co-axial with the threaded ends.

In the first series of tests, specimens were carefully machined, and were screwed into oiled spherical holders which were normally free to adjust themselves, but which had been wedged at the top and bottom to produce a definite but unknown amount of eccentricity. Careful stress-strain measurements were made on six gauge lines spaced 60° apart around the circumference. From the stress-strain curves obtained, it was shown that the effect of eccentricity was pronounced at first, producing considerable compression on one side, but that the compression was changed to tension at about one-third of the ultimate stress. While the side which was in initial compression never developed as large tensile strains, as did the side which was initially in tension, a large part of the difference vanished, and this was brought about by a slight bending of the loading pins into which the specimens were screwed, and to an equalising deformation of the threads of the specimen. That the relief of eccentricity resulted from the flexibility of the loading pins was borne out by a number of tests in which the test specimens were loaded eccentrically through devices which could not bend. It was also found that in tests, which were loaded eccentrically, and then carried to failure with an axial load, the ultimate strength was approximately equal to the values obtained by initially axial loads.

In the second series of tests, six bars, 12 in. long, were cut into three specimens each, machined and tested. The specimens were prepared with eccentricities of 0·0 in., 0·06 in. and 0·11 in., one from each bar. These were tested with screw threads in parallel planes, the specimens being screwed for a length of 1 in. into loading pins, which were clamped rigidly into the heads of the testing machine, so that there was practically no flexibility. The results of the tests are given in Table I, and show a reduction in strength of 4·1% and 14·9% for eccentricities of 0·06 in. and 0·11 in., respectively. Since the stress on the outermost fibre would be doubled with an eccentricity of 6·05 in., it would appear that the material was not sufficiently uniform to give satisfactory results, or else there was actually much less eccentricity present than was assumed to exist.

TABLE I.

ULTIMATE STRENGTH OF SPECIMENS LOADED WITH VARIOUS ECCENTRICITIES. LB. PER SQ. IN.

No.	Eccentricity in Inches		
	0·0	0·06	0·11
1	30,100	24,500	21,200
2	21,900	25,700	21,400
3	24,100	25,600	19,600
4	25,600	24,400	22,700
5	26,800	24,200	23,600
6	25,100	23,100	22,300
Average	25,600	24,600	21,800

Tests were also made in which a different method was used to produce eccentric loads. The specimens were rectangular in cross section, being obtained from the walls of a cast-iron water pipe, and the loads were applied by means of pins through holes drilled at certain distances from the central axis of the specimens. Two of the five specimens tested had holes on the axis, and three had holes 0·15 in. from the axis. The eccentricity of 0·15 in. was chosen because, theoretically, it gave no stress on one side of the specimen, and double the average stress on the other side. Two gauge lines with a gauge length of 4 in. were placed on each face of each specimen, and strain measurements, made with a removable strain gauge on each of the four gauge lines, showed the average tensile strain and failure to be 0·0060 in. and 0·0068 in. per in., respectively, for the axial- and eccentrically-loaded specimens. A summary of the results obtained is given in Table II.

TABLE II.

STRESS-STRAIN DATA FOR AXIAL AND ECCENTRIC LOADS.

No.	Axial Load		Eccentric Load	
	Maximum Strain In./In.	Ultimate Strength Lb. per Sq. in.	Maximum Strain In./In.	Ultimate Strength Lb. per Sq. in.
1	0·0053	20,500	—	—
2	0·0068	19,900	—	—
3	—	—	0·0059	14,000
4	—	—	0·0075	15,000
5	—	—	0·0070	15,500
Average	0·0060	22,200	0·0068	14,800

From these results it is seen that there is a reduction of 27% in the ultimate stress under eccentric loading. Theoretically, there should be a reduction of 50%, but this value would be modified slightly by the deflection of the eccentrically-loaded specimen by 6% or 7% to 43%. The difference between 43% and 27% must be due to a difference in the strength of the material, or to the distribution of stress across the specimen, or both. The assumed distribution of stress is shown in Fig. 1(a), but the actual distribution is probably more like that shown in Fig. 1(b).

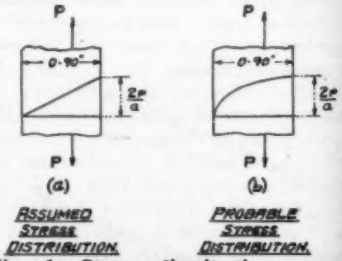

Fig. 1.—Stress distribution across specimen.

In general, the data obtained show a substantial reduction in the loads carried by grey cast-iron bars under eccentric loads, as compared with the loads carried under axial loads, but the exact amount of this reduction is affected by so many conditions that it cannot be stated with certainty.

From *The Foundry*, 1945, vol. 73, No. 7, pp. 94-95, 180 and 182.